MARCH, 1958

JOURNAL OF THE IRON AND STEEL INSTITUTE

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VOL. 188 PART 3

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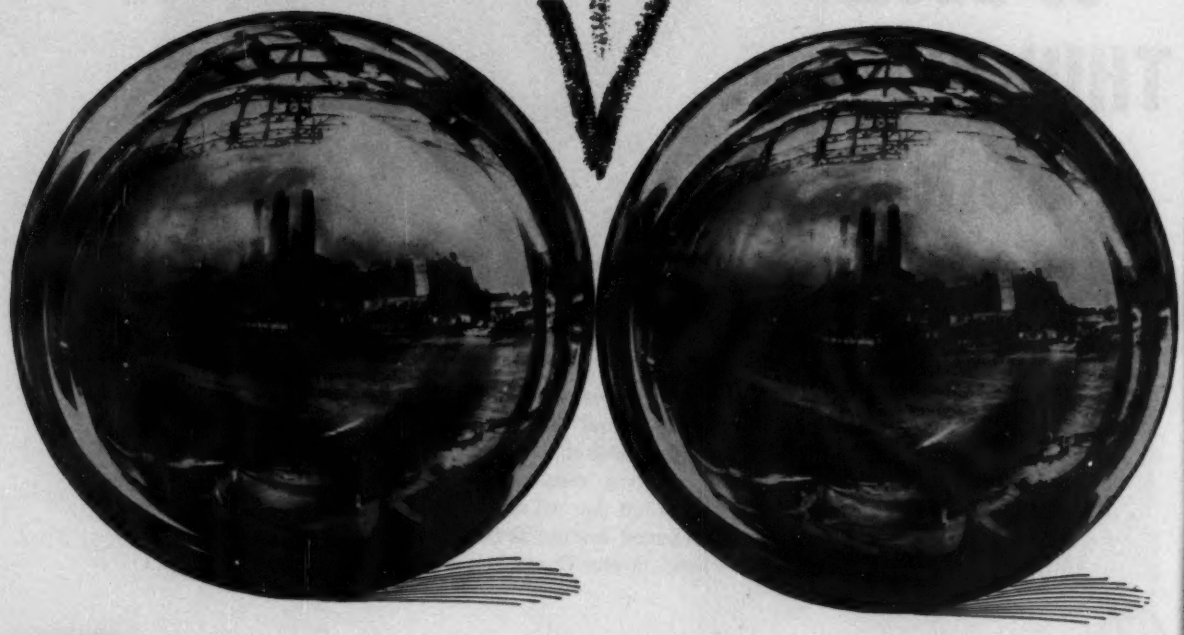

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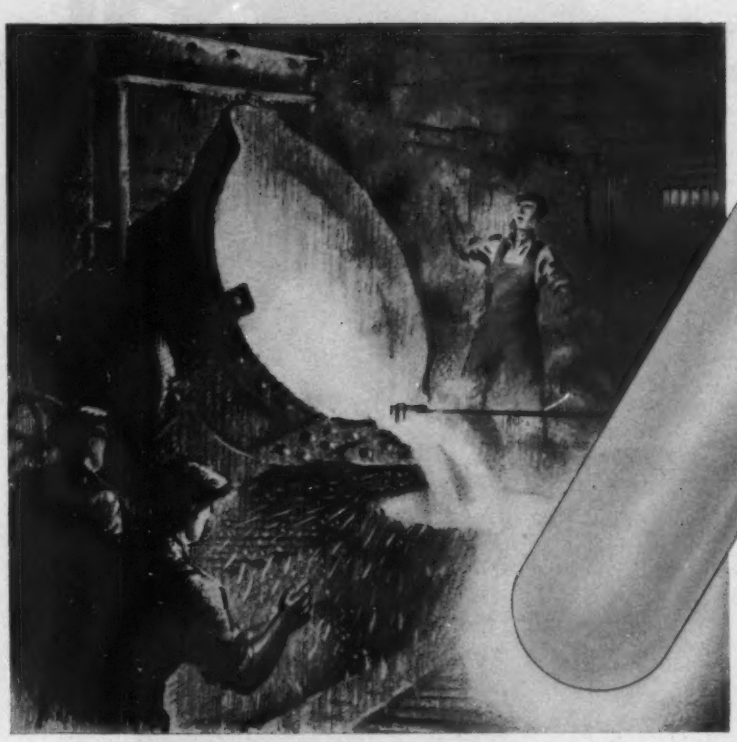

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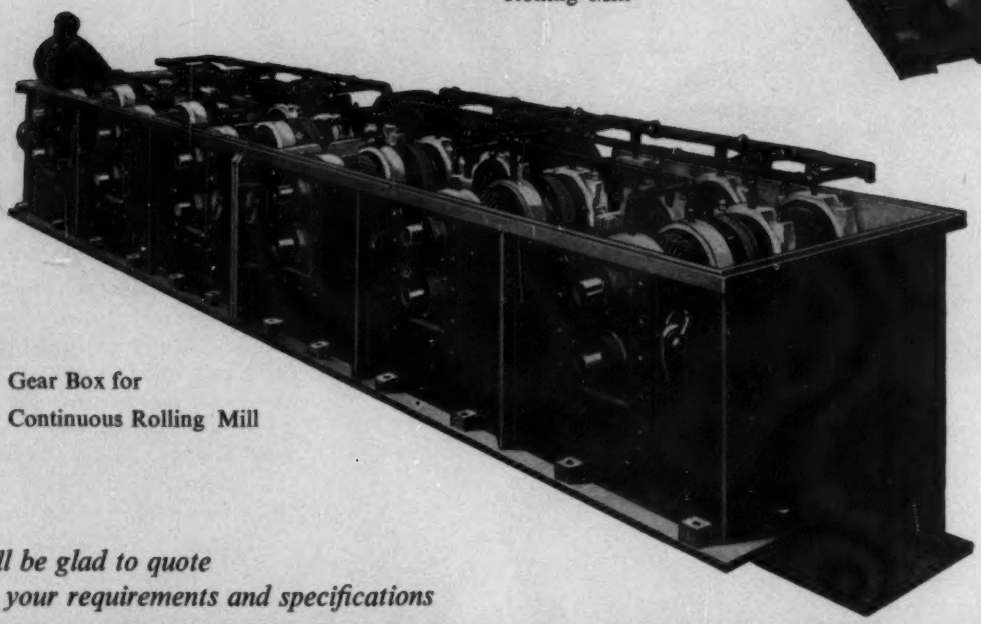

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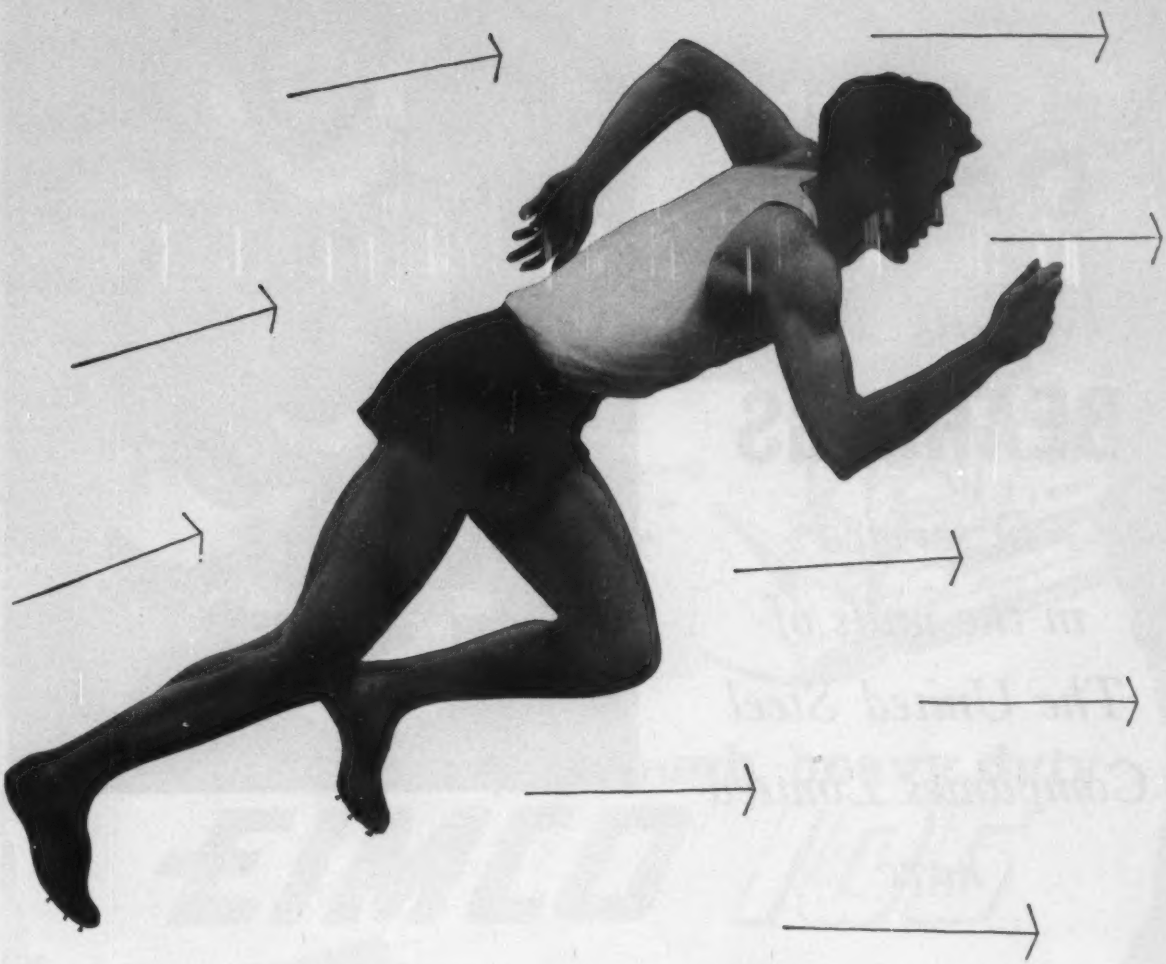

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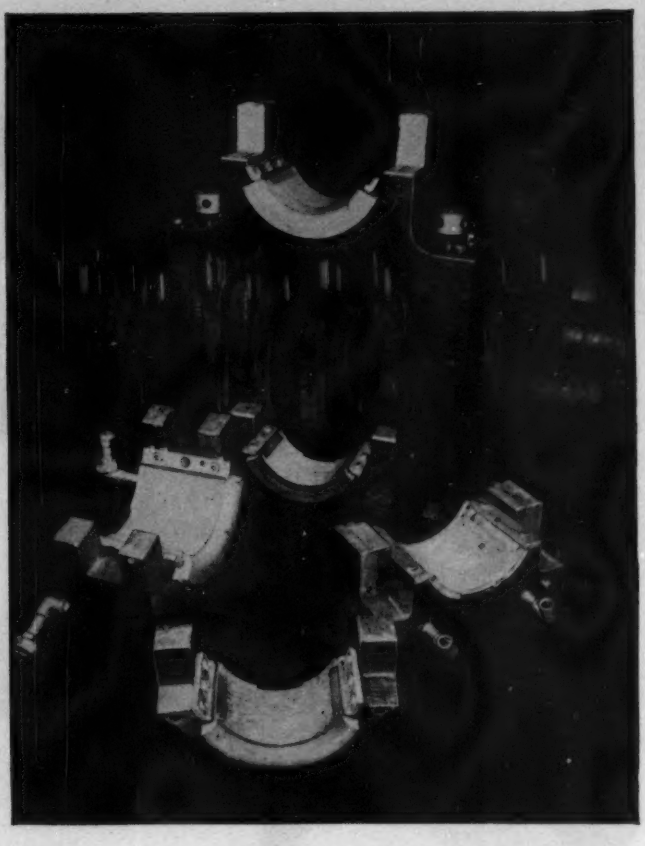

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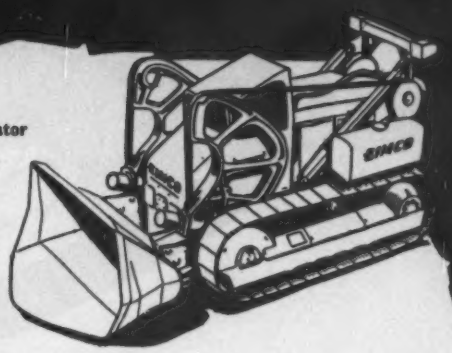

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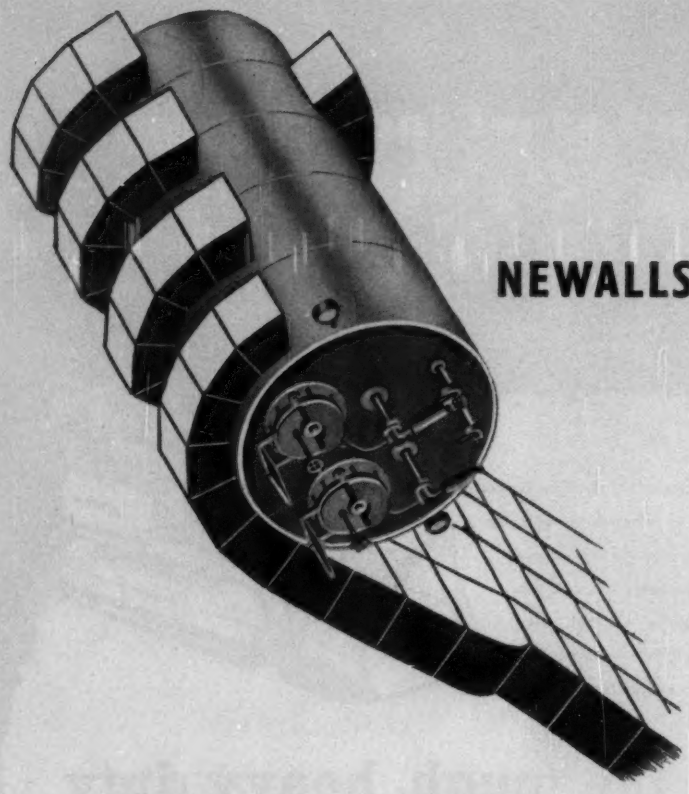

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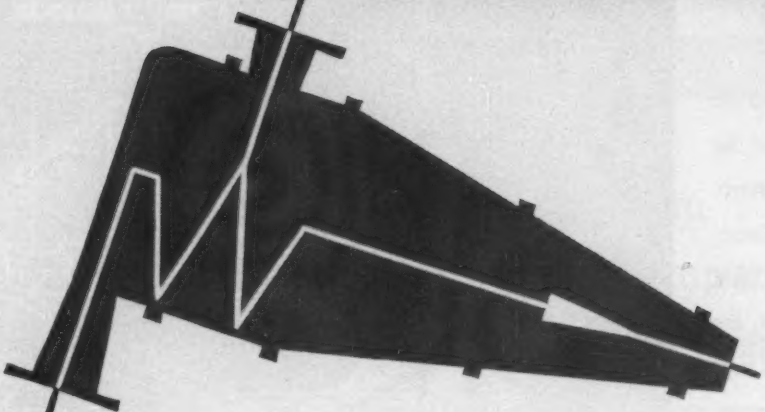

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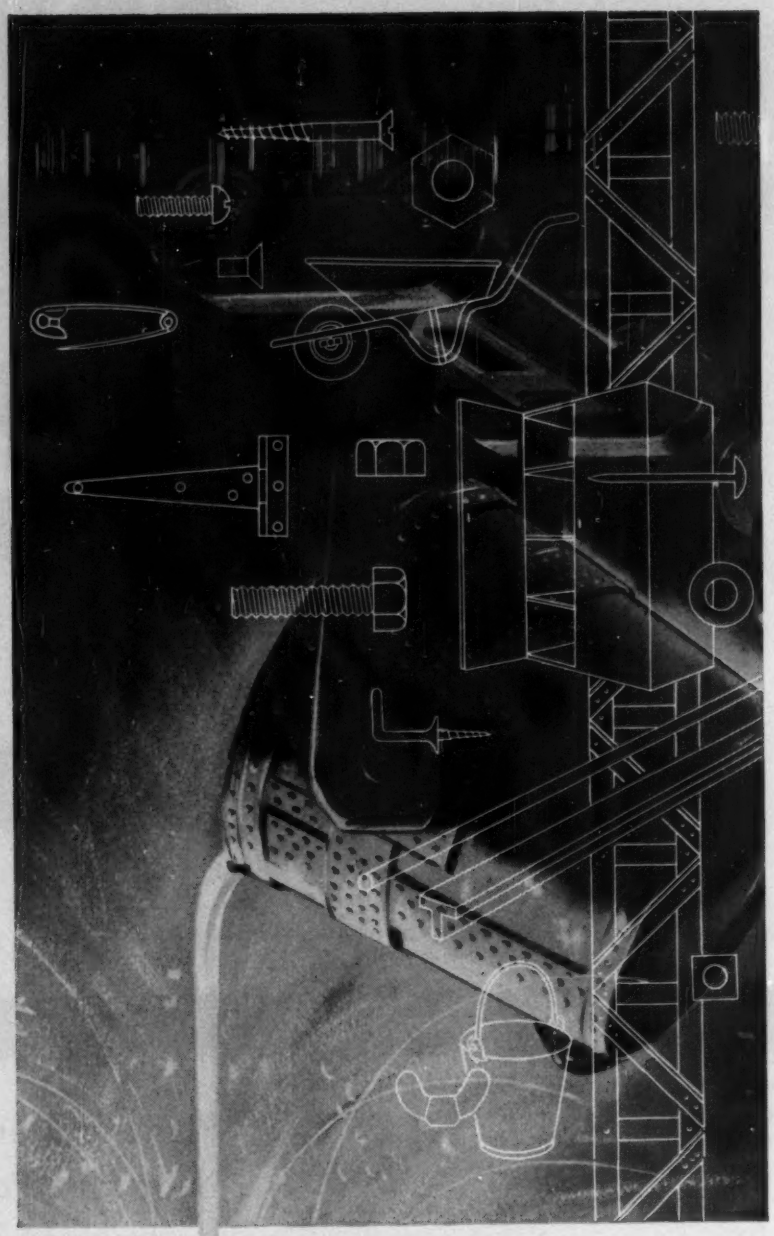

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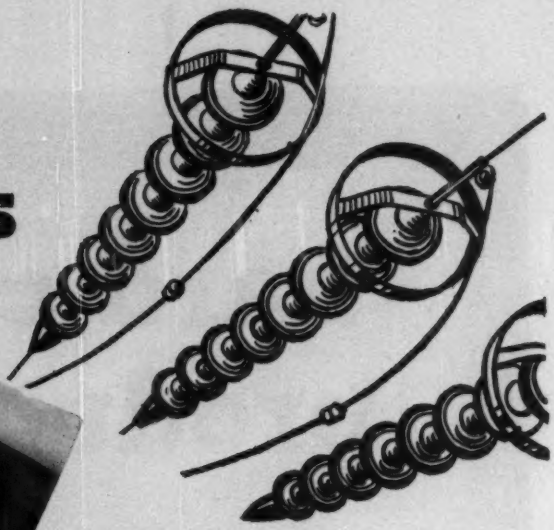

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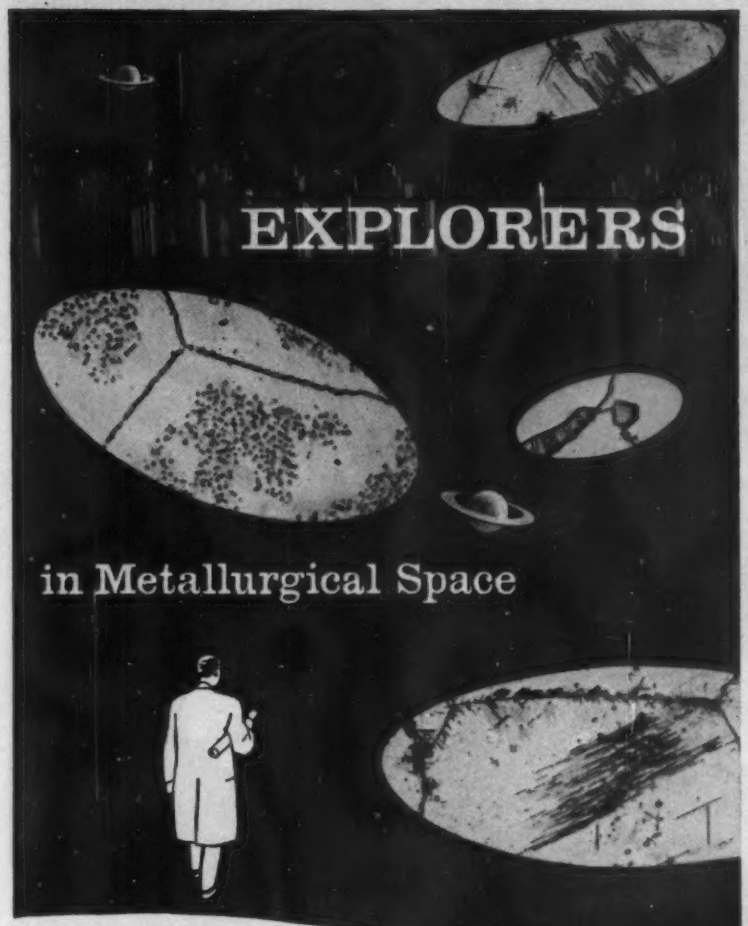

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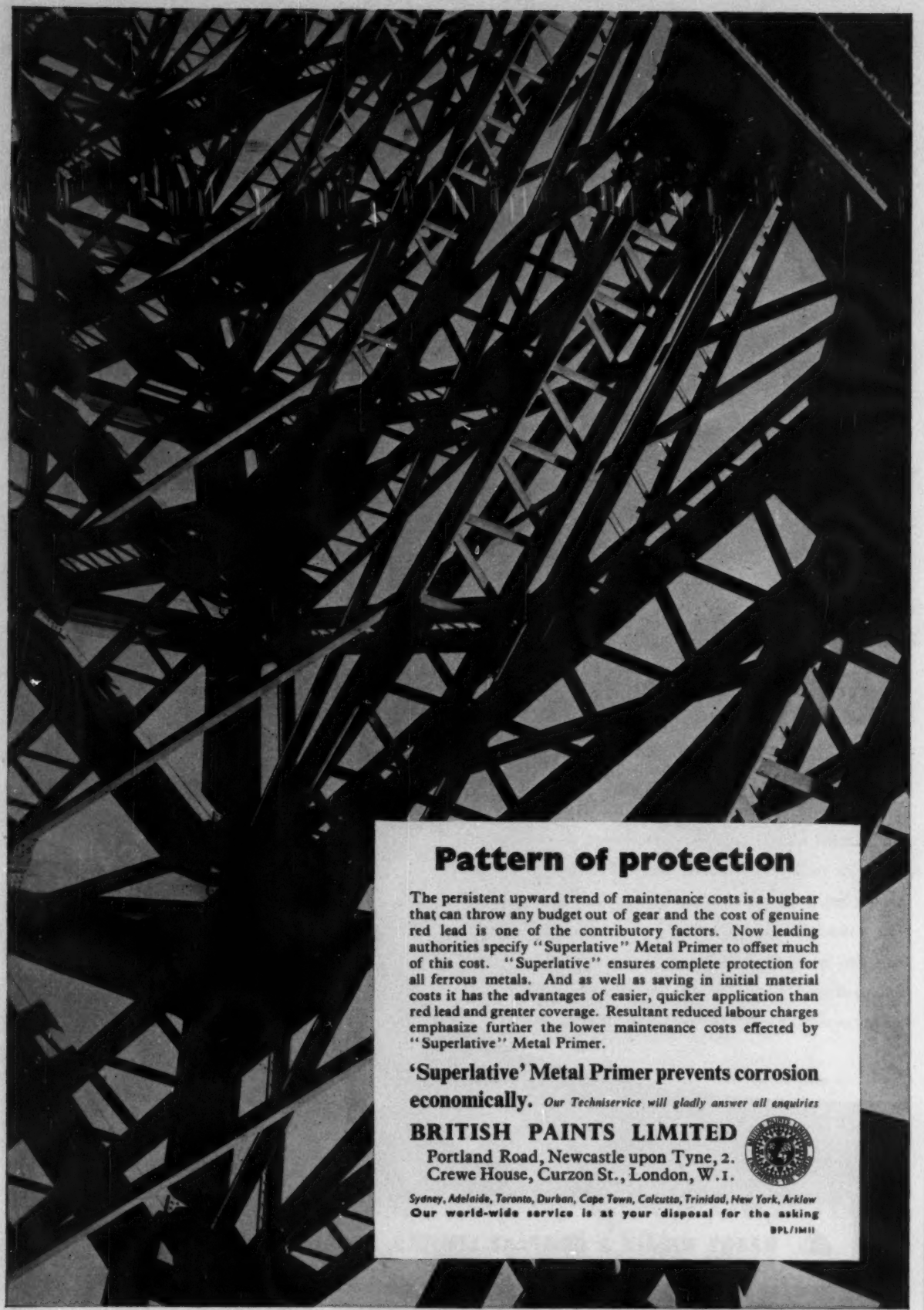
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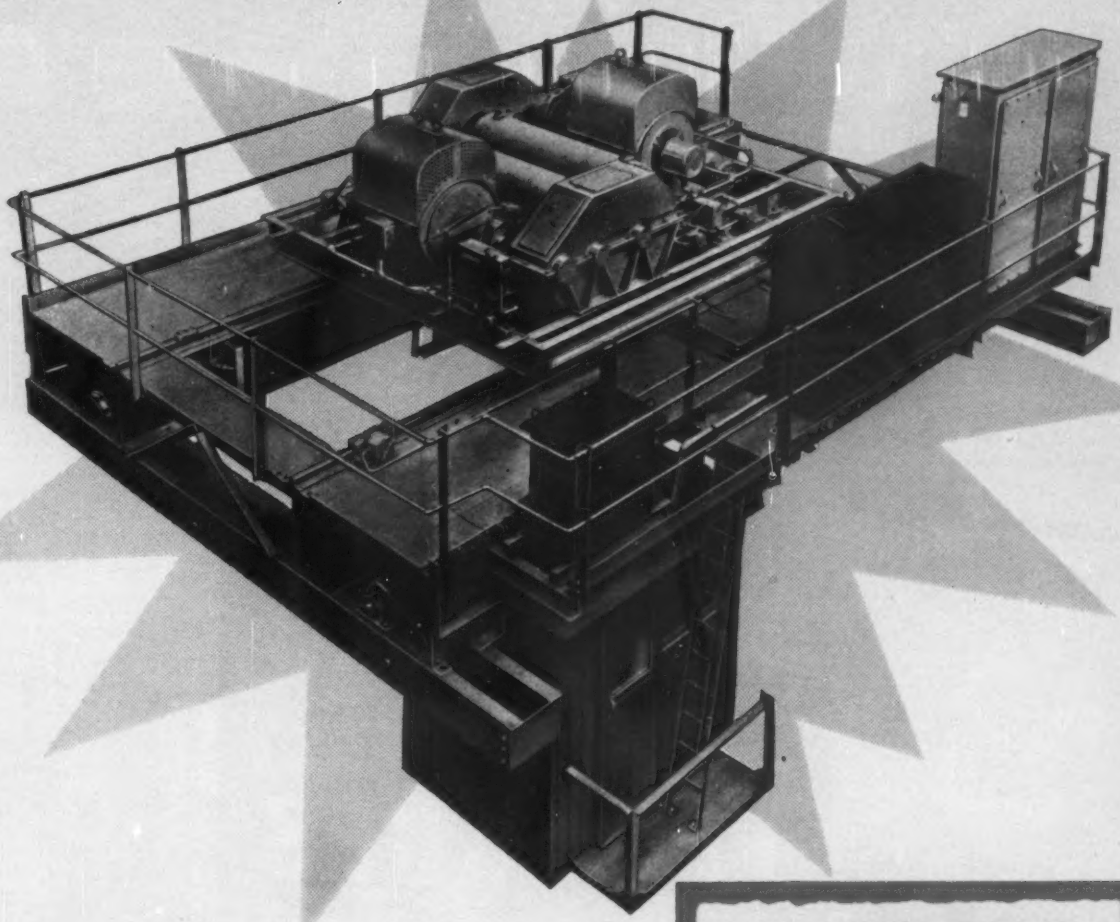

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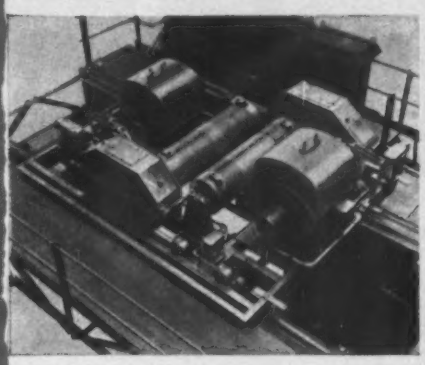

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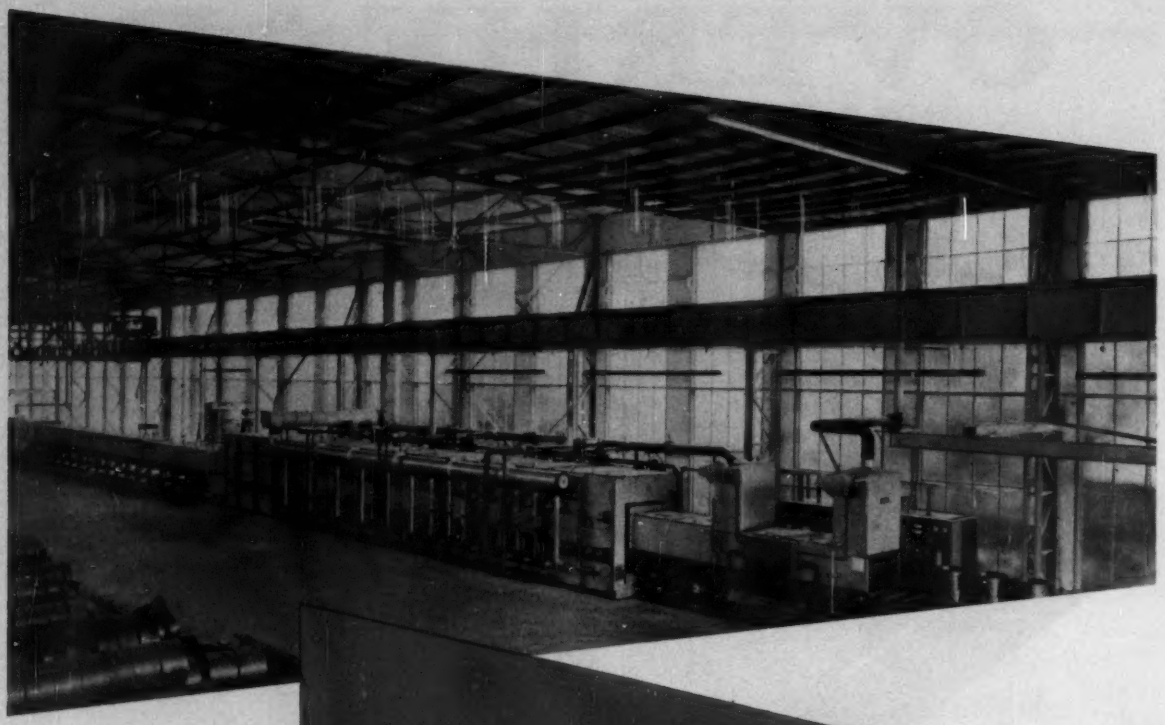

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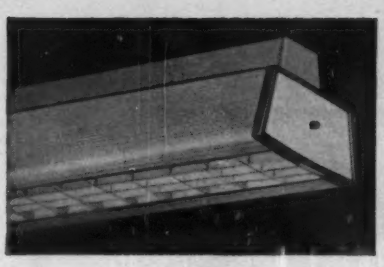

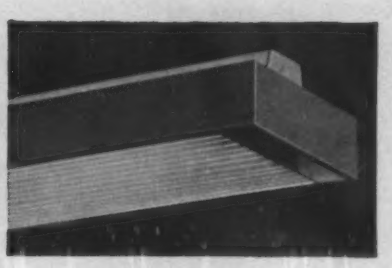

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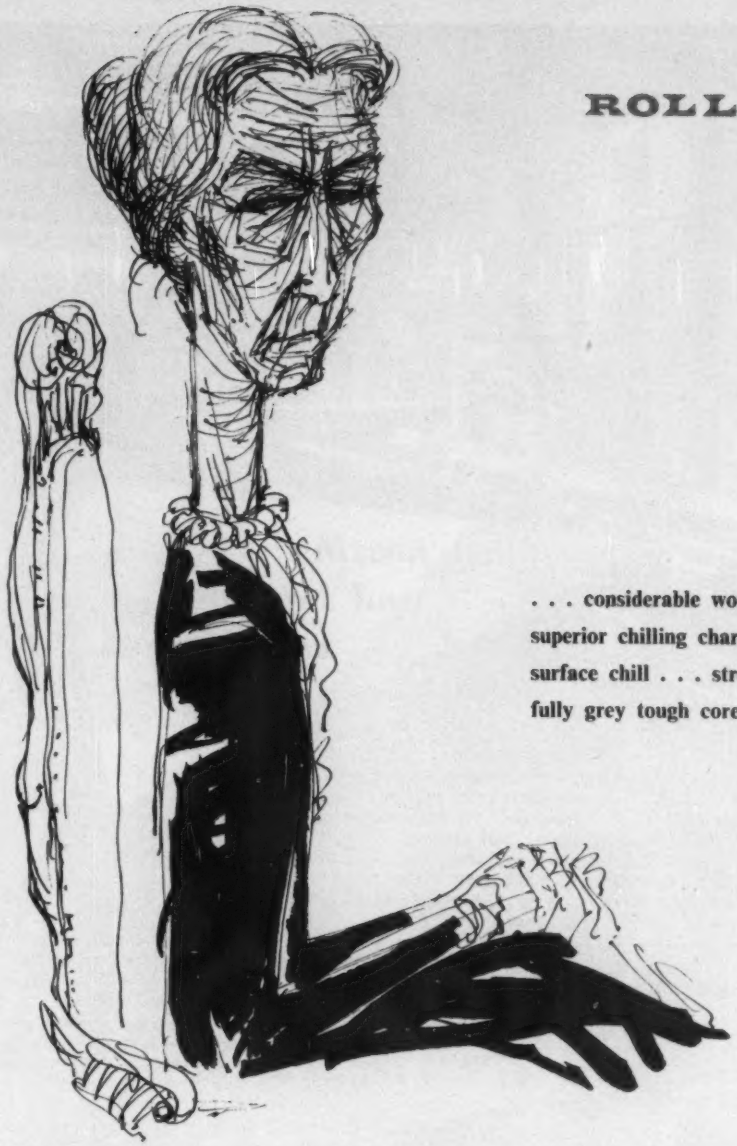

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We specialise in the design and erection of the following furnaces:
OPEN HEARTH, ROTARY HEARTHS complete with Automatic Charge and Discharge Machines. SOAKING PITS, DIRECT FIRED ANNEALING FURNACES, CONTINUOUS AUTOMATIC PUSHER AND ROLLER HEARTH type furnaces for Heat Treatment.
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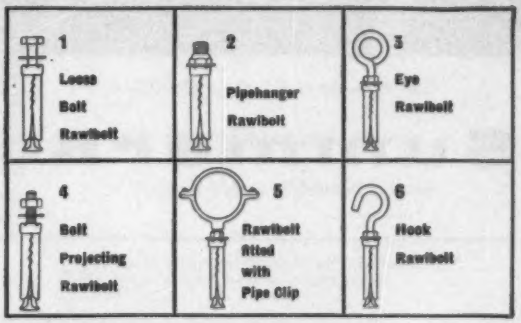
Rawlbolt charts showing full range of sizes and types with illustrated literature available on receipt of letter-heading or business card.

6 HEADS are better than one!

You never need to improvise with Rawlbolts. There are types exactly right for every bolt-fixing job, No. 5, for example, for the support of pipe-work, and No. 2 for use with Pipe-hangers and other gas fittings. Hook and Eye Rawlbolts (Nos. 3 and 6) have many uses, such as providing anchorage for guy ropes and cables, supporting suspended ceilings, etc. There's the Loose Bolt Rawlbolt (No. 1), inserted after the machine is slid into position, and the Bolt Projecting Type (No. 4), ideal for wall-fixings.

Rawlbolts give you bolt-fixings of enormous strength in a fraction of the time taken by any other method.



Every bolt-fixing job is far quicker with

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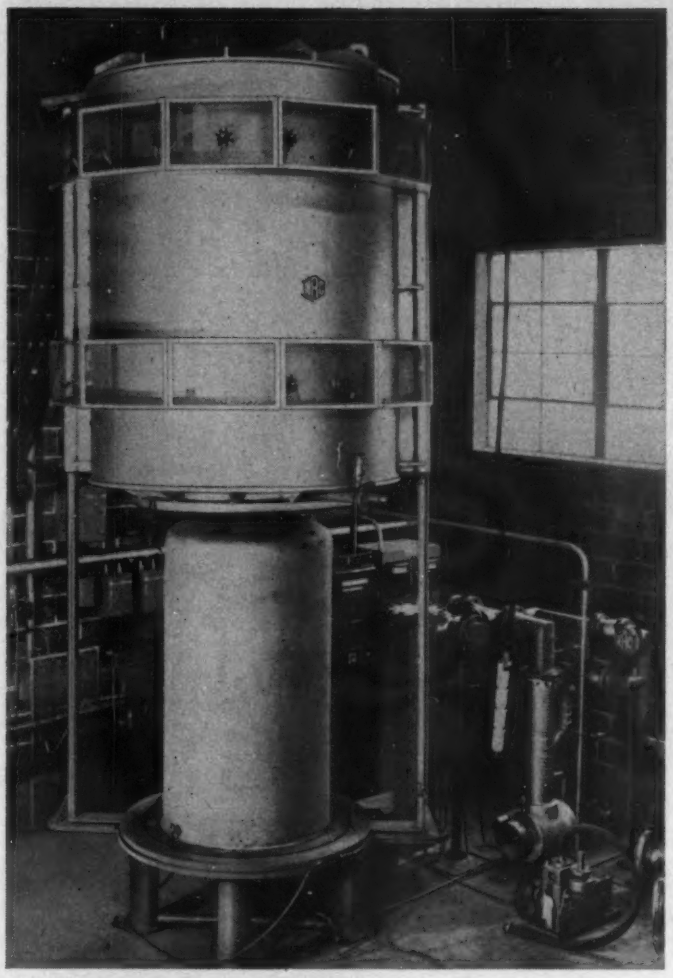

This model has a uniform hot zone 36" dia x 36" high and an ultimate vacuum of less than 1 micron.

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* NRC *is the trademark of the National Research Corporation, registered in the United States Patent Office.*

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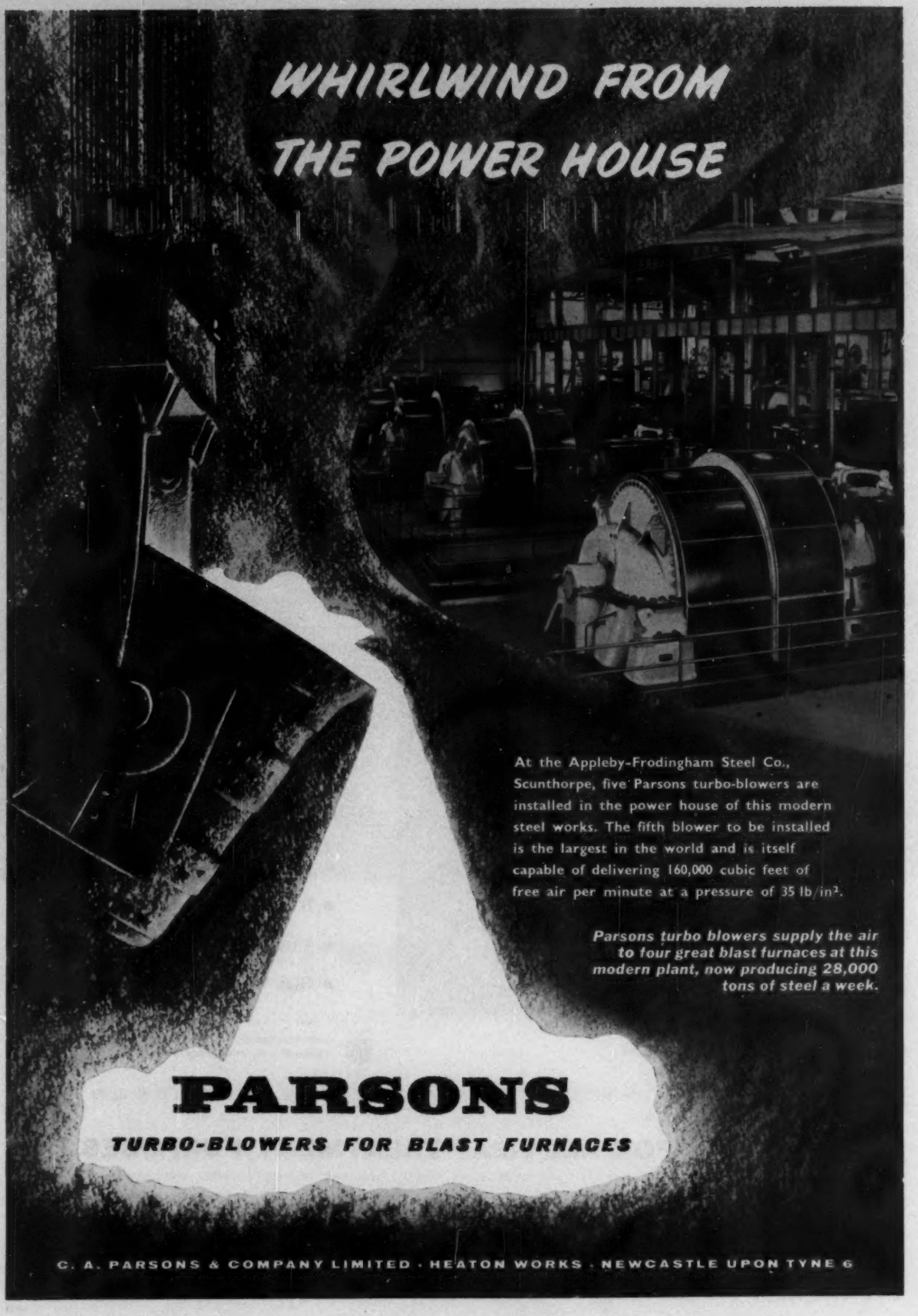
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THE POWER HOUSE
At the Appleby-Frodingham Steel Co., Scunthorpe, five Parsons turbo-blowers are installed in the power house of this modern steel works. The fifth blower to be installed is the largest in the world and is itself capable of delivering 160,000 cubic feet of free air per minute at a pressure of 35 lb/in².
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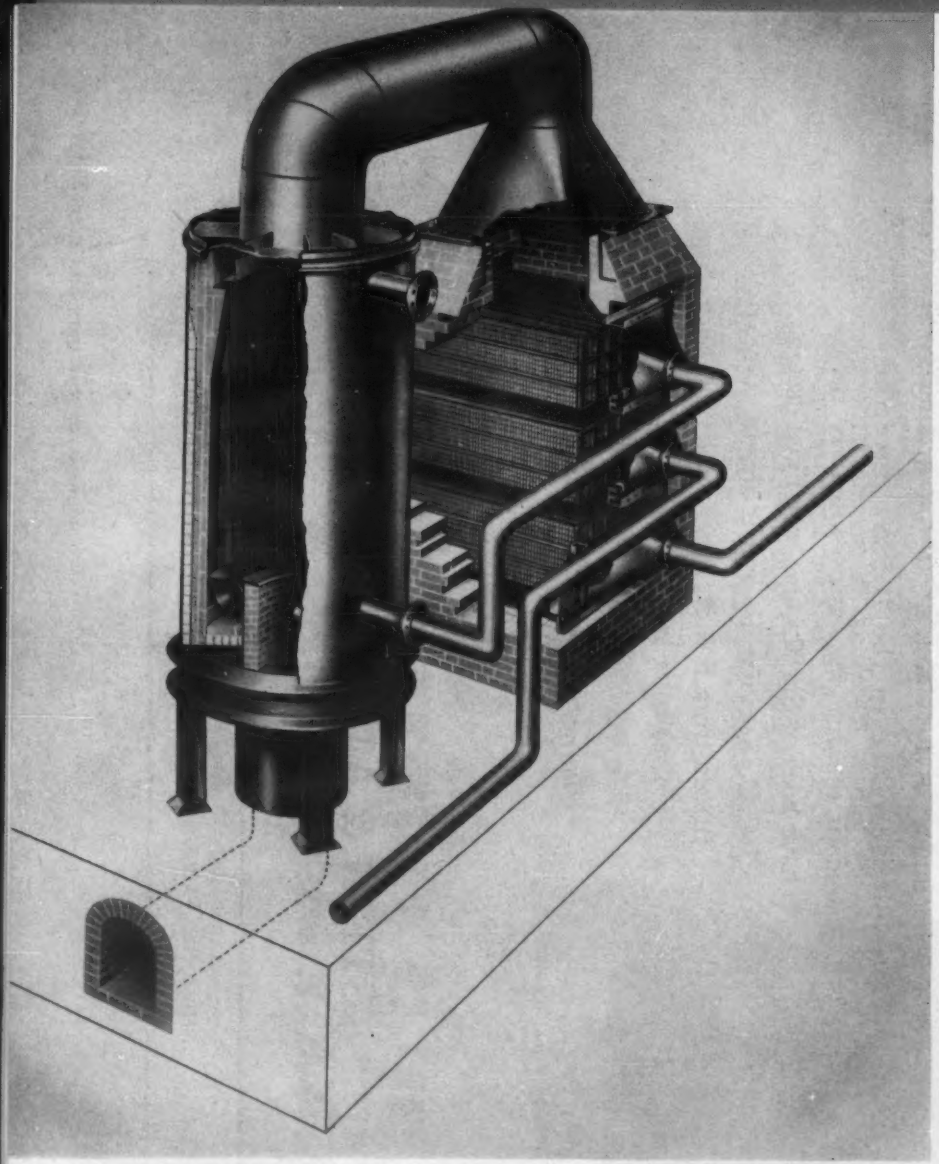

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new radiant type preheater

NEWTON CHAMBERS have now completed their agreement with Industri-kemiska A.B. of Stockholm, for the manufacture of INKA radiant type air and gas preheaters throughout Britain and the Commonwealth.

Newton Chambers have been specialists for many years in the manufacture of all kinds of fuel economy equipment, and with the long awaited INKA can now offer equipment for preheating air, steam and combustible gases, etc., up to 850°C. by flue gases, or direct firing. Other advantages will be:

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. . . and our American associates, Surface Combustion Corporation, have been selected for this installation of 48 soaking pits for the Gary Works of the United States Steel Corporation.

This follows an order for 18 pits for the same Works in 1955-56. S.C.C. have had orders for more than 200 pits since the beginning of 1955 and, of all the soaking pits installed in the U.S.A. since the end of the last war, **approximately 90%** have been of Surface One-way Fired design. S.C.C. service is available in Great Britain from S & A.

We believe that these facts will be of profound interest to British Steelworks managements who may be considering future development plans.

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THE 'DUPAR' Type 'M' Contactors are designed to give trouble-free service and long life under the most arduous conditions. They incorporate features which practically eliminate contact 'bounce' and its effect on contact life through pitting and burning, and provide highly efficient 'blowout' on both alternating and direct current circuits.

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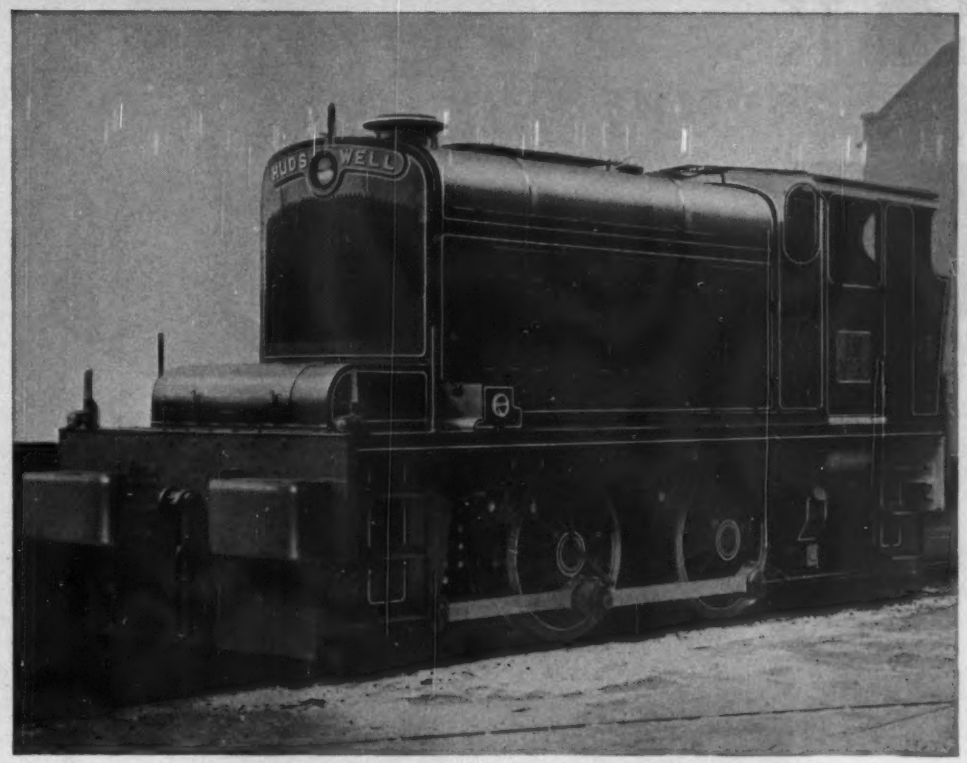

For English Steel Corporation Ltd...

A 300 B.H.P. 0-4-0 Type Diesel Locomotive Standard Gauge weighing 43 tons recently delivered to the River Don Works of English Steel Corporation Limited, Sheffield. Fitted with a Crossley 5 Cylinder two stroke non-reversing engine running at 900 R.P.M. Vulcan Sinclair Scoop Controlled Fluid Coupling with S.S.S. Power Flow Gear Box: 3 speeds forward and reverse: Auxiliaries driven through a Vulcan Sinclair Coupling.

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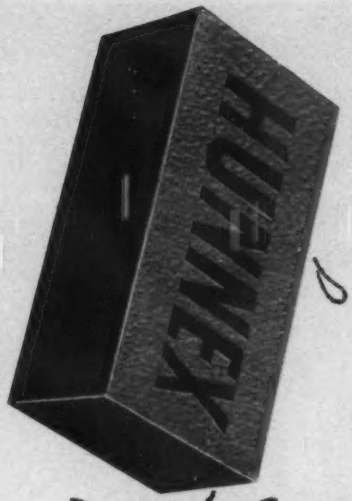

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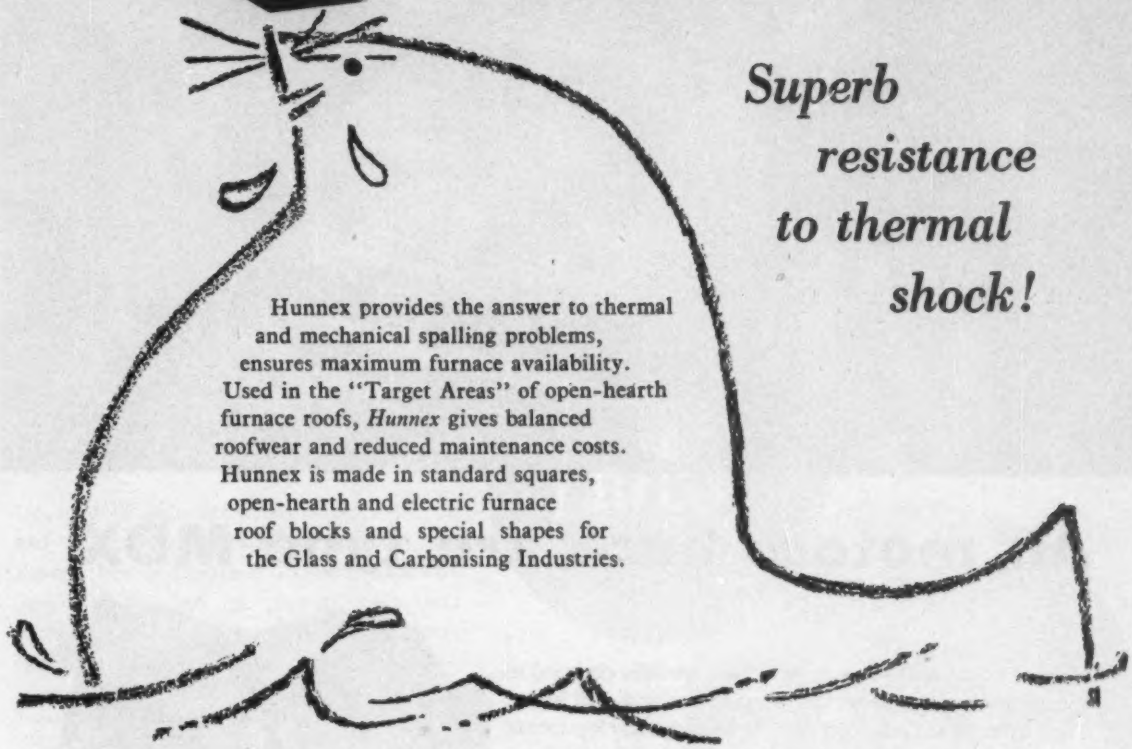
Hunnex provides the answer to thermal and mechanical spalling problems, ensures maximum furnace availability. Used in the "Target Areas" of open-hearth furnace roofs, *Hunnex* gives balanced roofwear and reduced maintenance costs. Hunnex is made in standard squares, open-hearth and electric furnace roof blocks and special shapes for the Glass and Carbonising Industries.



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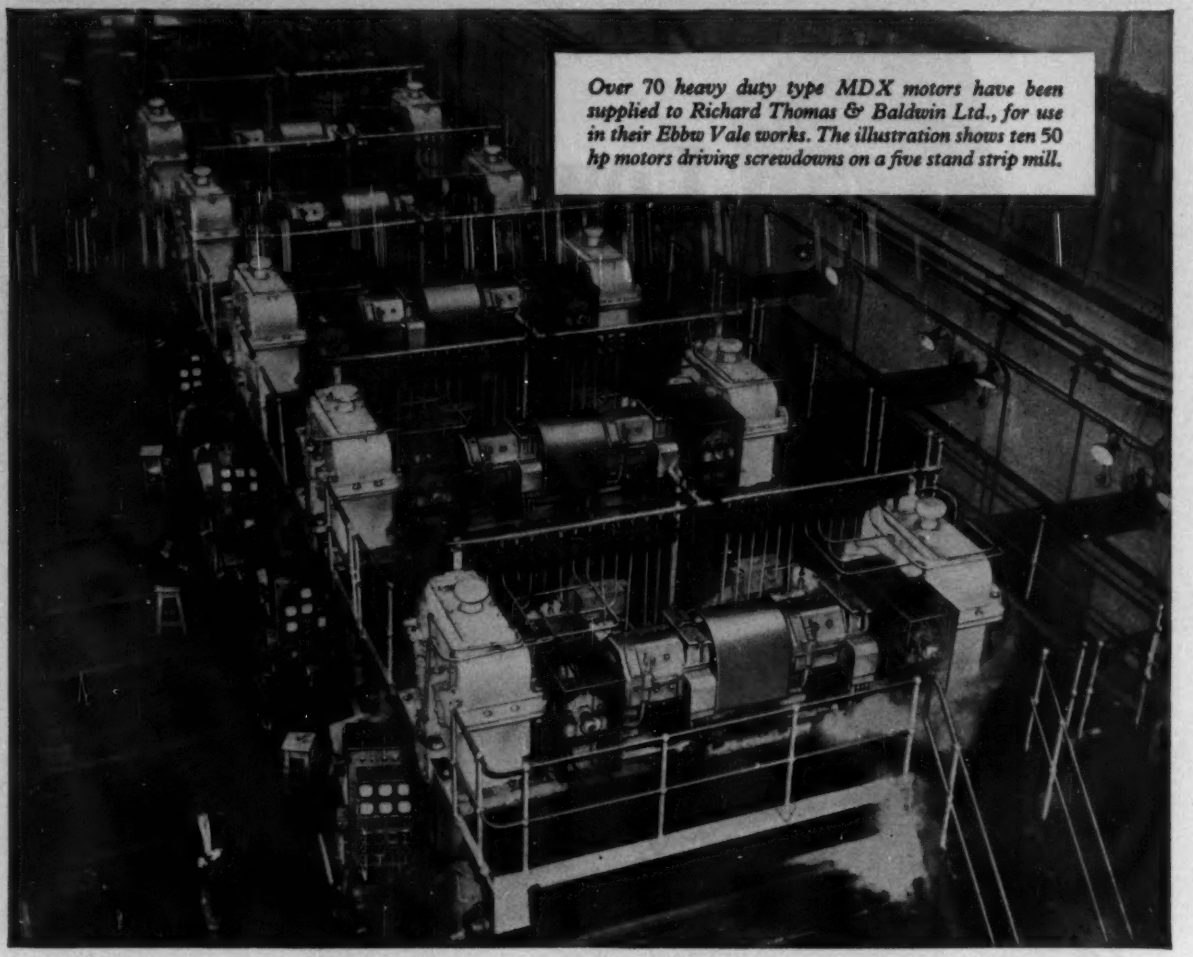
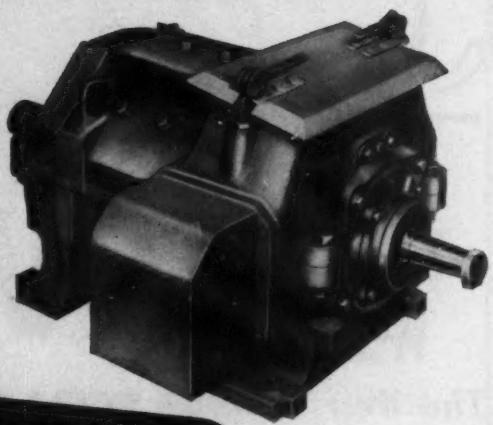


Over 70 heavy duty type MDX motors have been supplied to Richard Thomas & Baldwin Ltd., for use in their Ebbw Vale works. The illustration shows ten 50 hp motors driving screwdowns on a five stand strip mill.

All motors here are type MDX



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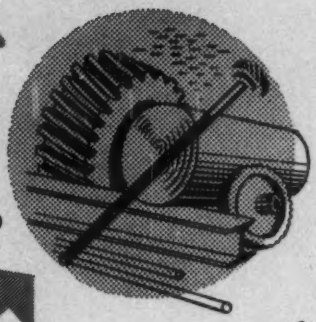
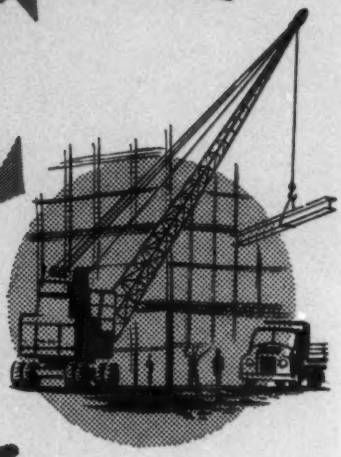


dismantling.....



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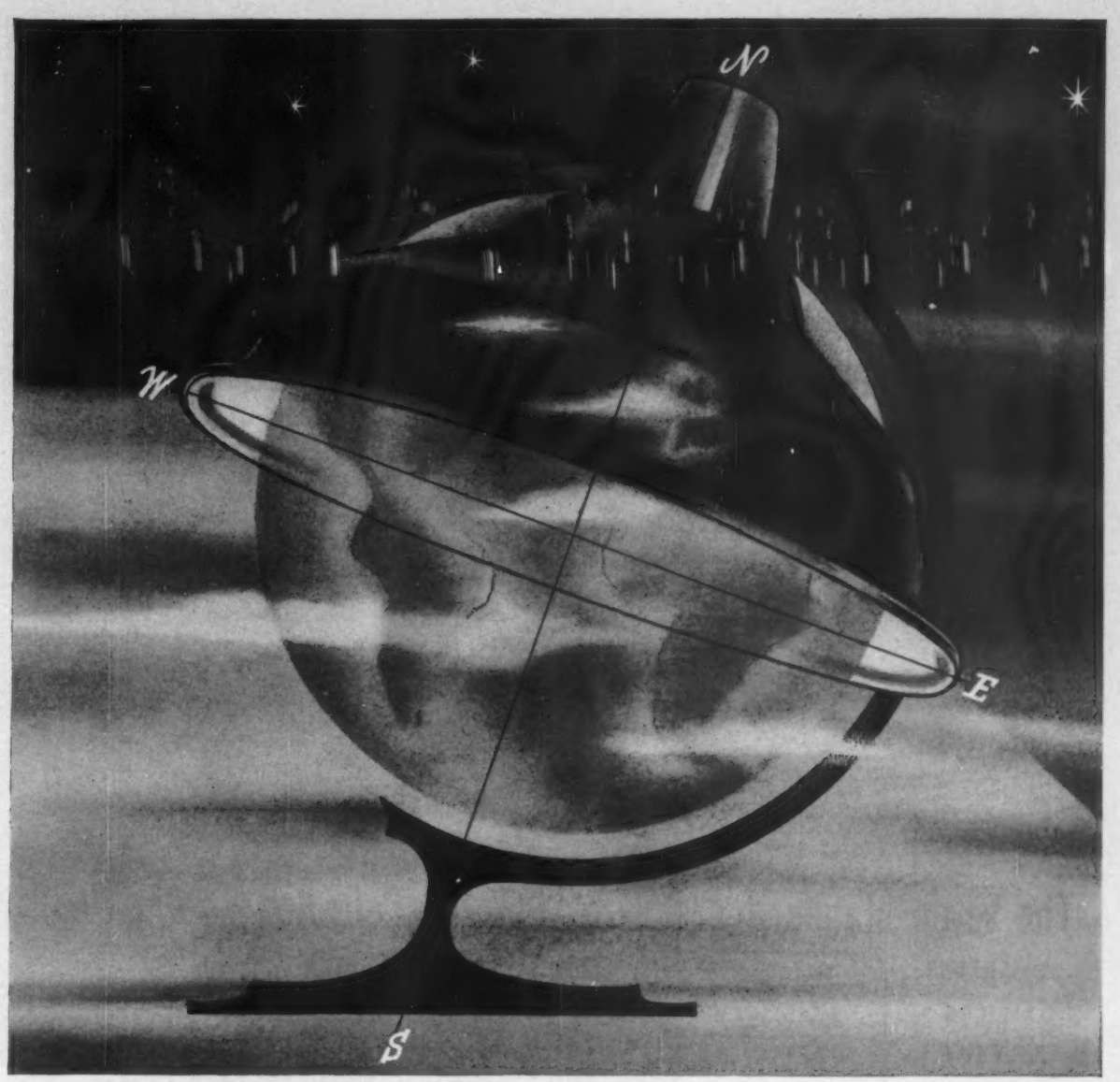

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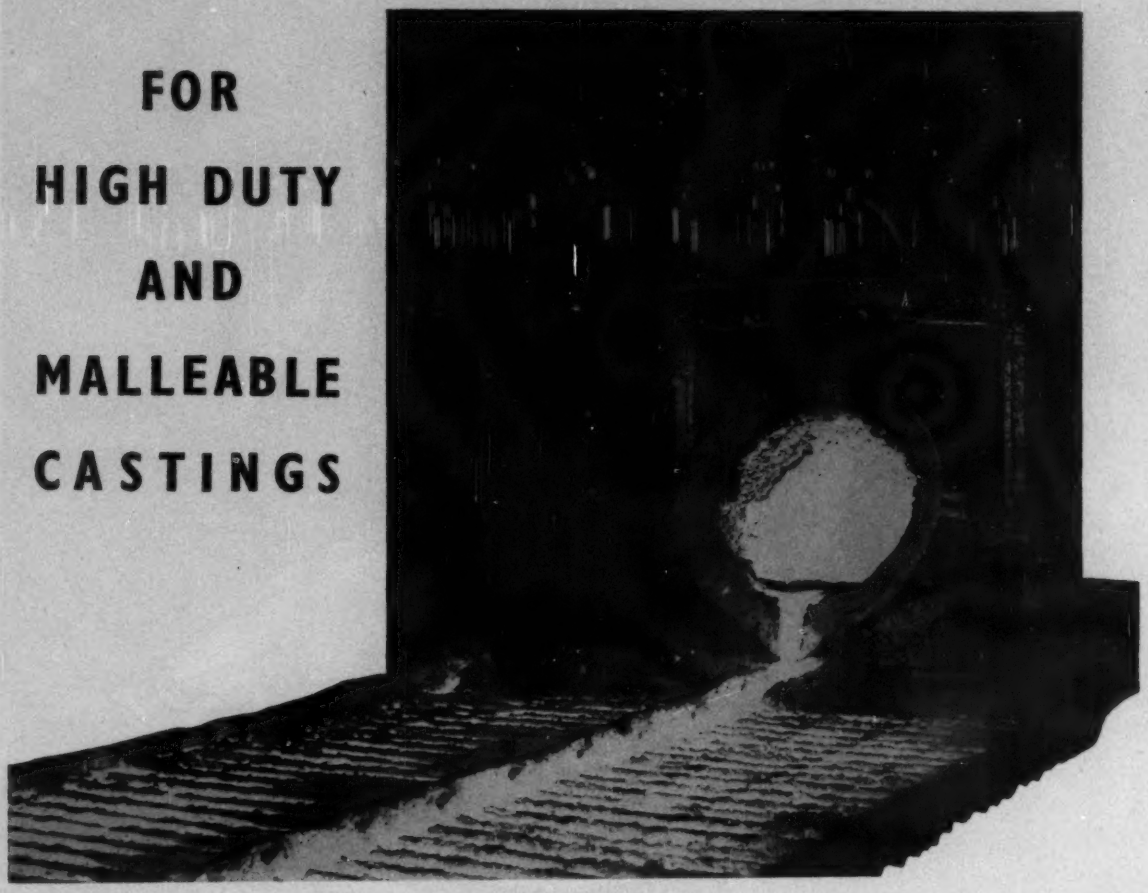

The seven standard grades of Stanton Dale Refined Pig Iron

GRADES:	A	B	C	D	E	F	G
SILICON %	1·9–2·1	1·4–1·6	0·9–1·1	2·25–2·5	0·9–1·1	1·4–1·6	1·9–2·1
SULPHUR %	0·06	0·07	0·08	0·05	0·08	0·07	0·06
PHOSPHORUS %	0·6–0·7	0·6–0·7	0·6–0·7	0·4–0·5	0·25	0·25	0·25
MANGANESE %	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5
TOTAL CARBON %	2·7–3·0				2·8–3·1		

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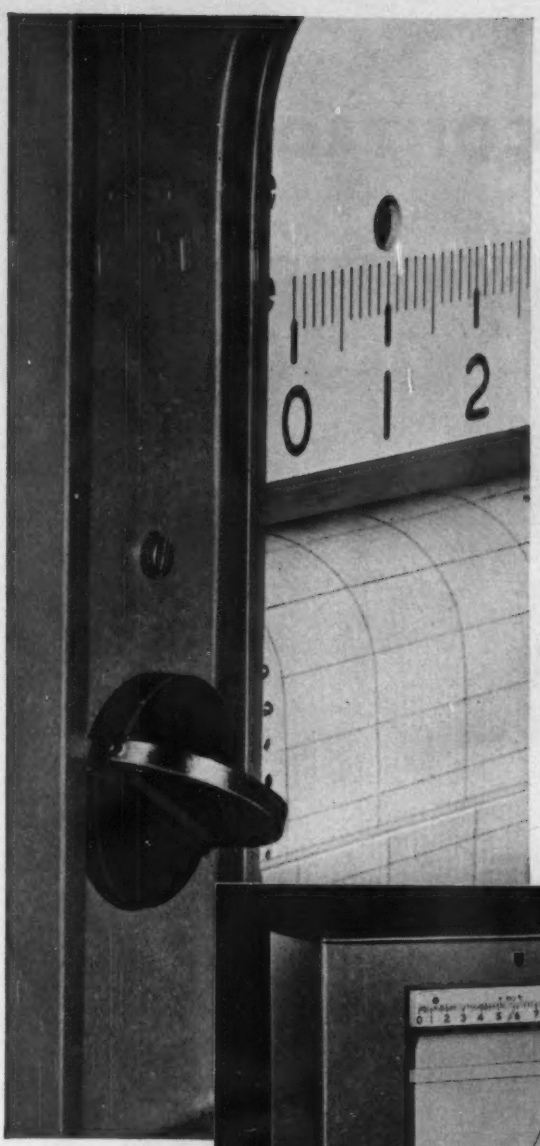

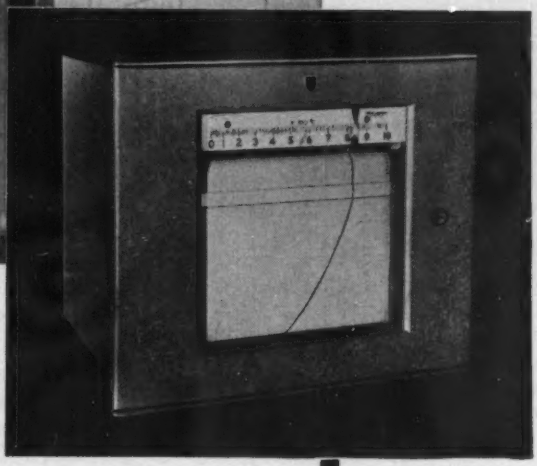

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(5) Installation on N.E. Coast accommodating 160 3½/4-ton ingots, fired with blast furnace gas and having fully automatic reversal equipment and air/gas ratio control.

(6) Installation in the Midlands of 4-hole plant accommodating 40 3/3½-ton ingots, fired with 100 per cent. cold blast furnace gas with automatic reversal equipment, gas/air ratio and furnace pressure control with completely sealed covers.

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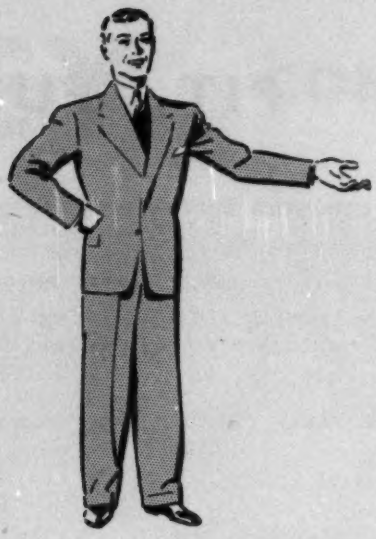

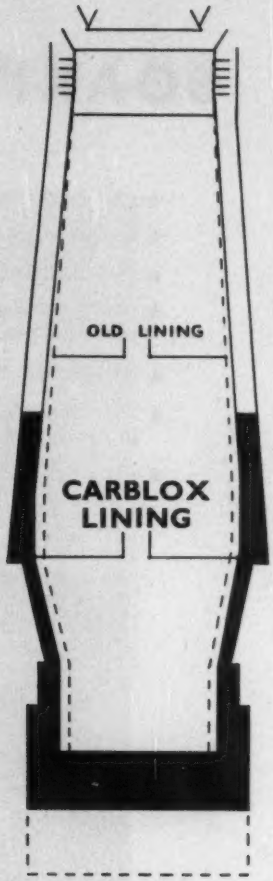

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For the
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More iron can be made in an existing blast furnace if a thinner lining is used to give greater furnace capacity. This can be achieved by using Carblox which has a much greater resistance to the attack of iron and slag than normal firebrick. The after contraction is low and its high thermal conductivity increases cooling efficiency. Ordinary firebrick fails at 2910°F under a load of only 50lb/sq. in. yet CARBLOX shows no deformation under a load of 2 tons/sq. in. CARBLOX is 2-3 times as resistant to abrasion and its mean cold crushing strength is about four times that of normal firebrick. Burden, molten slag and iron do not stick to it. CARBLOX must be used under reducing conditions.

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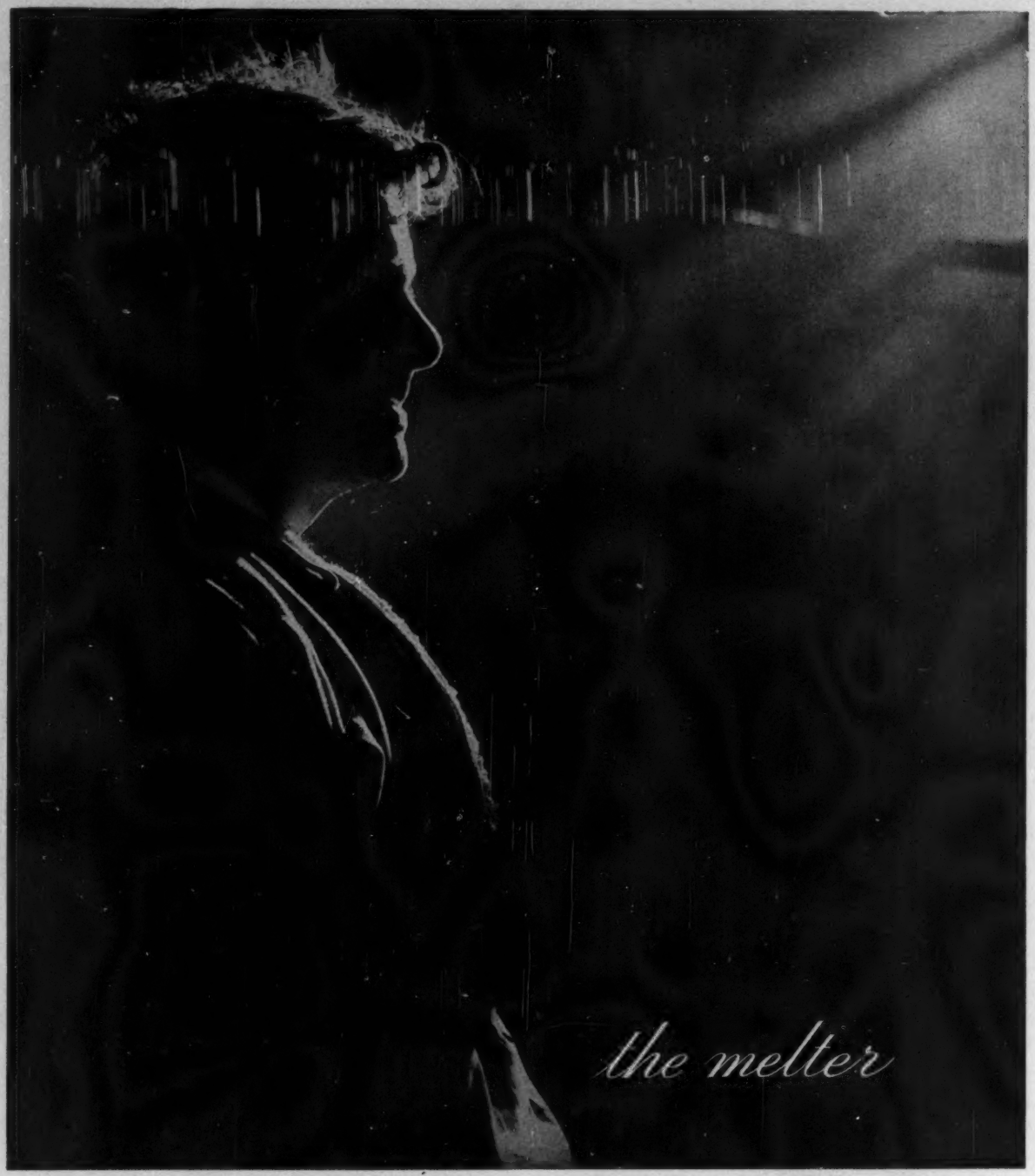

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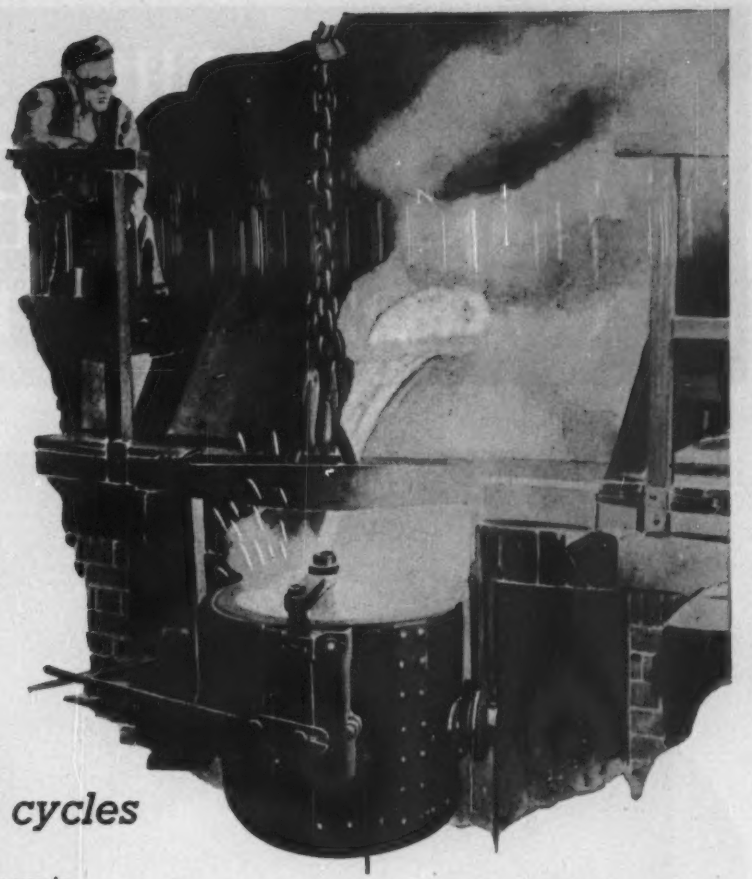

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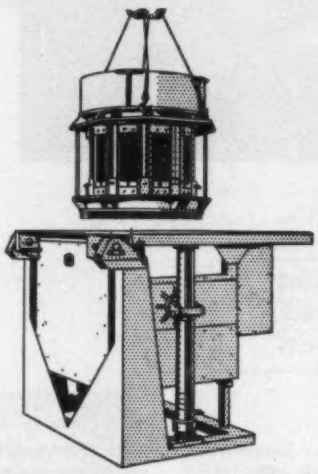

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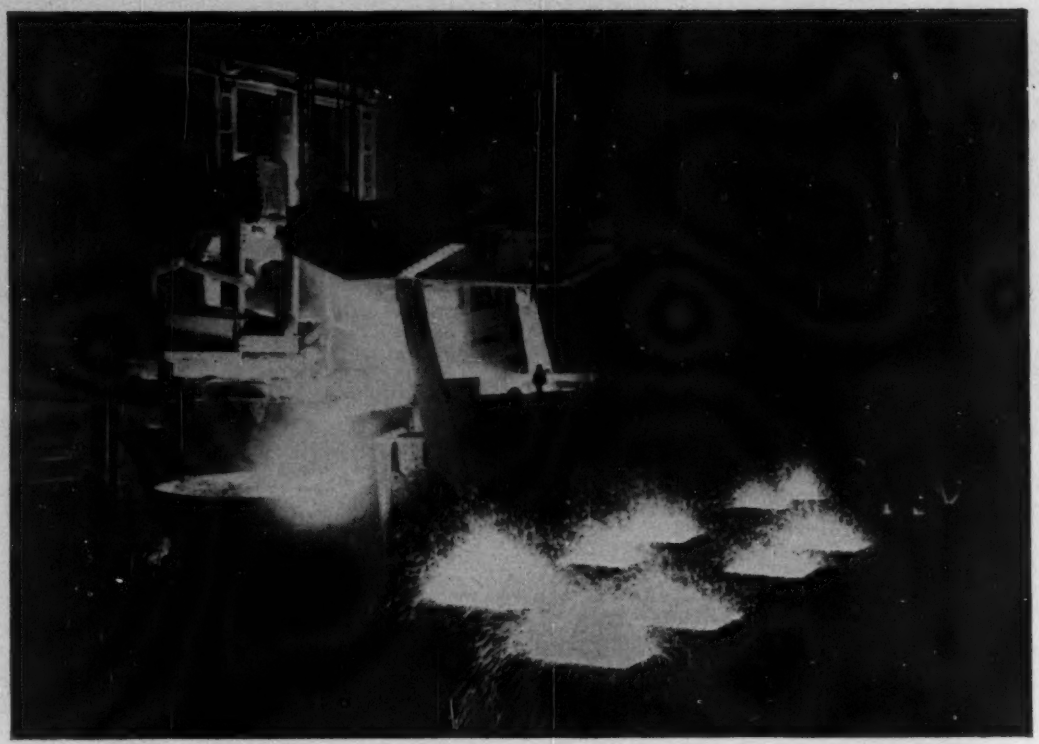

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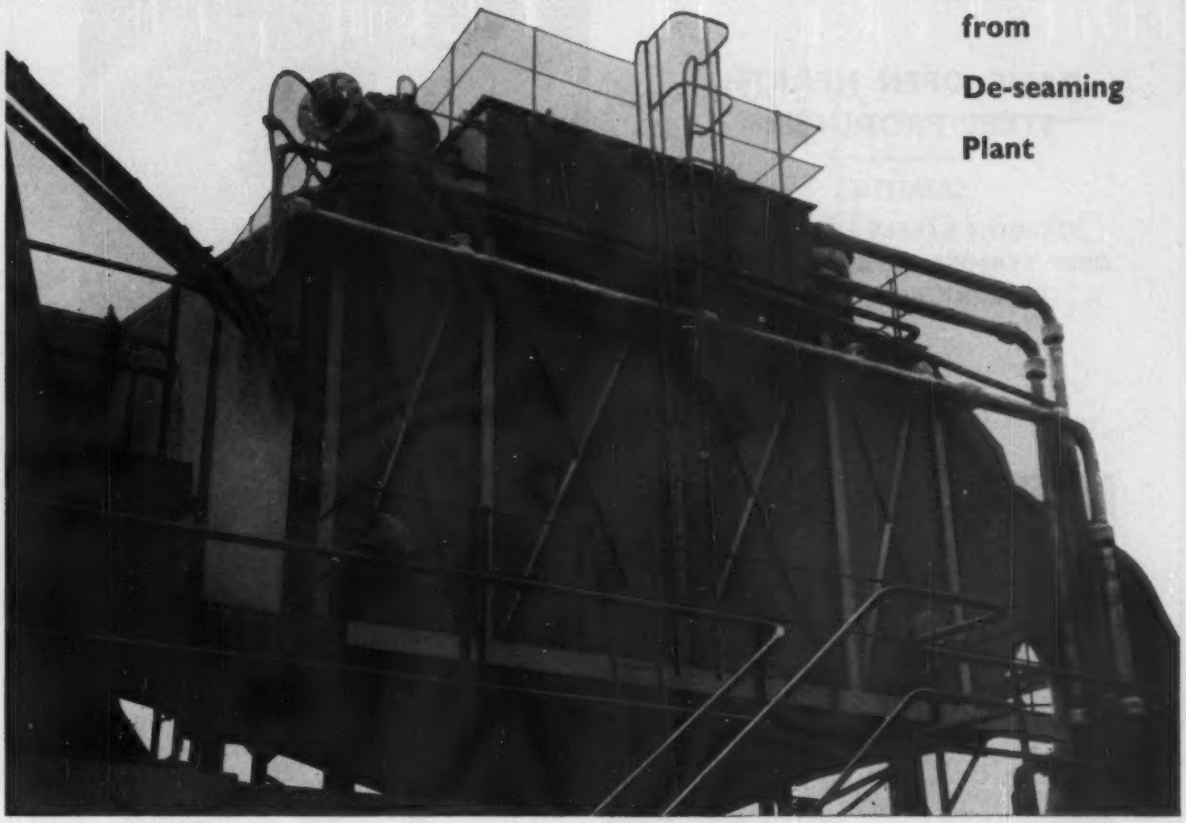

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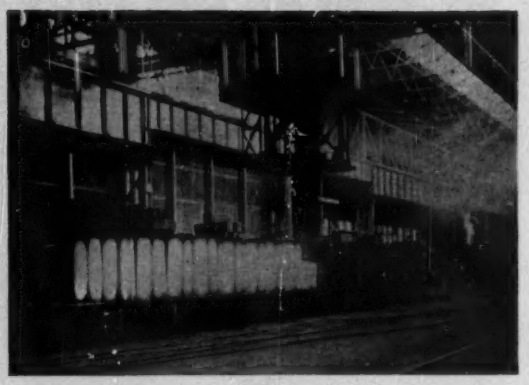

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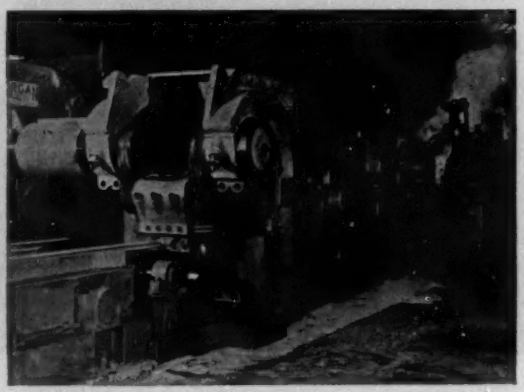

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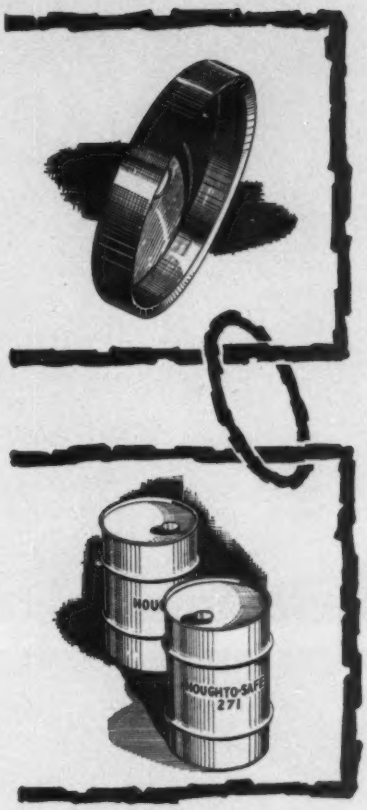

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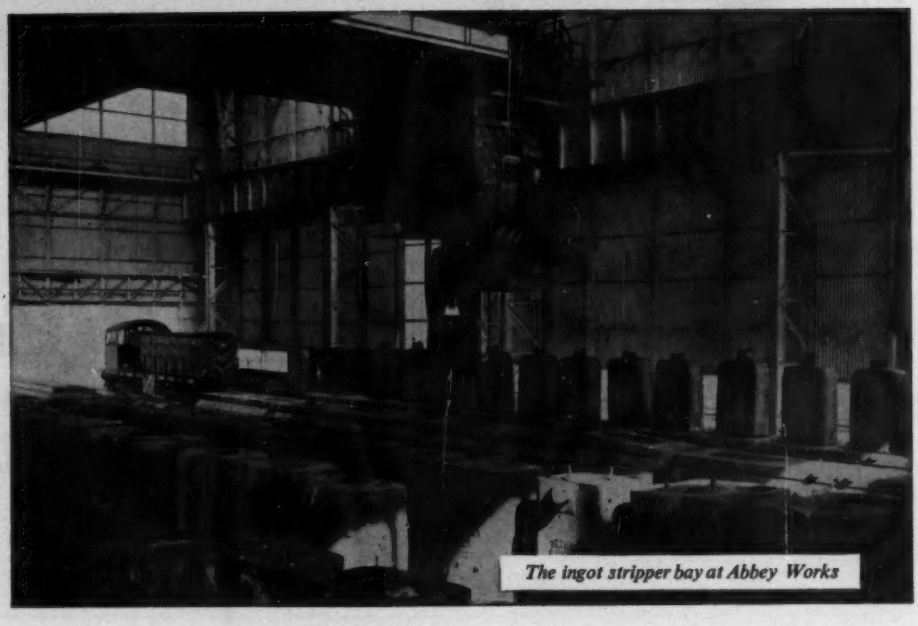

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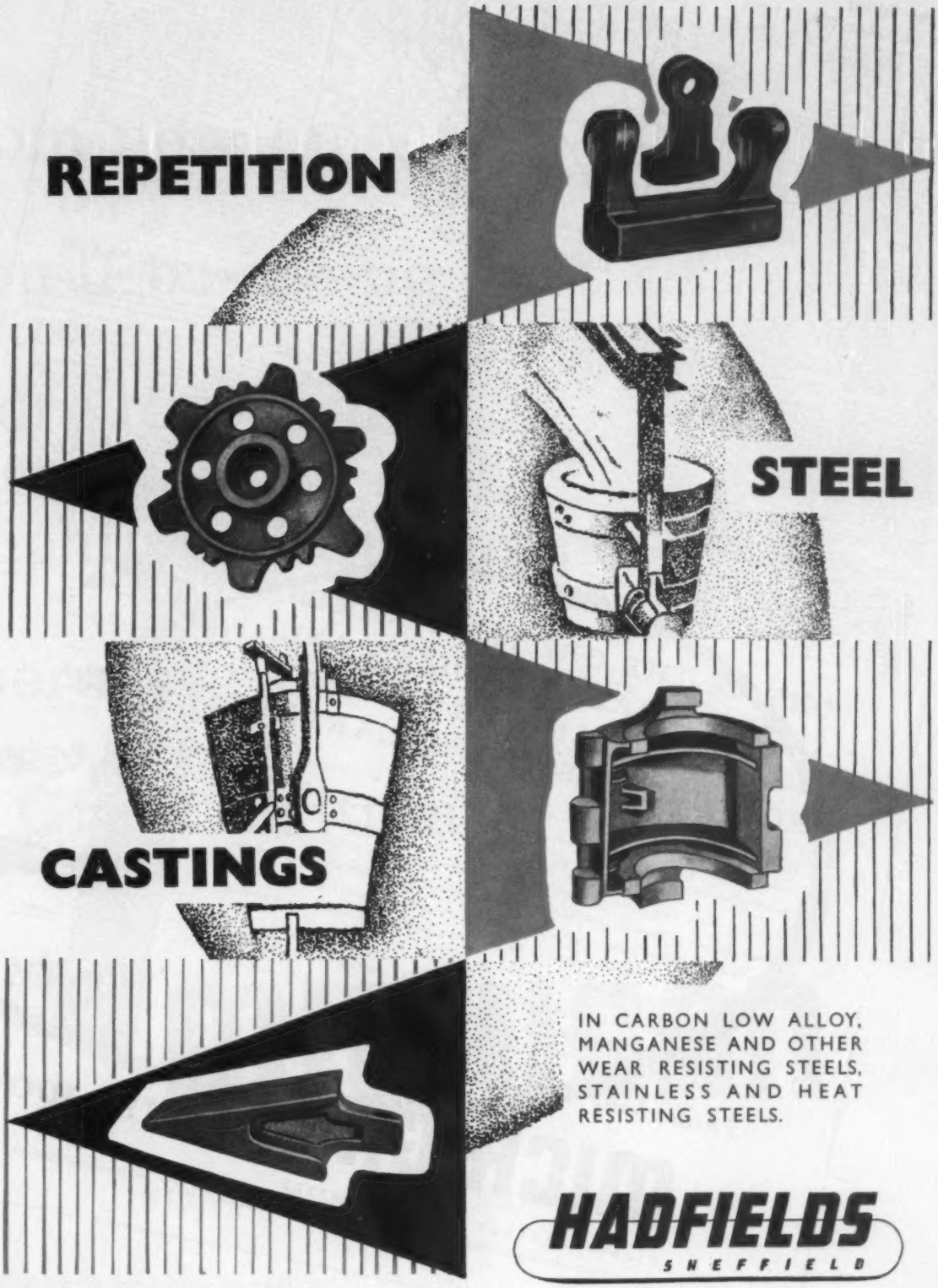

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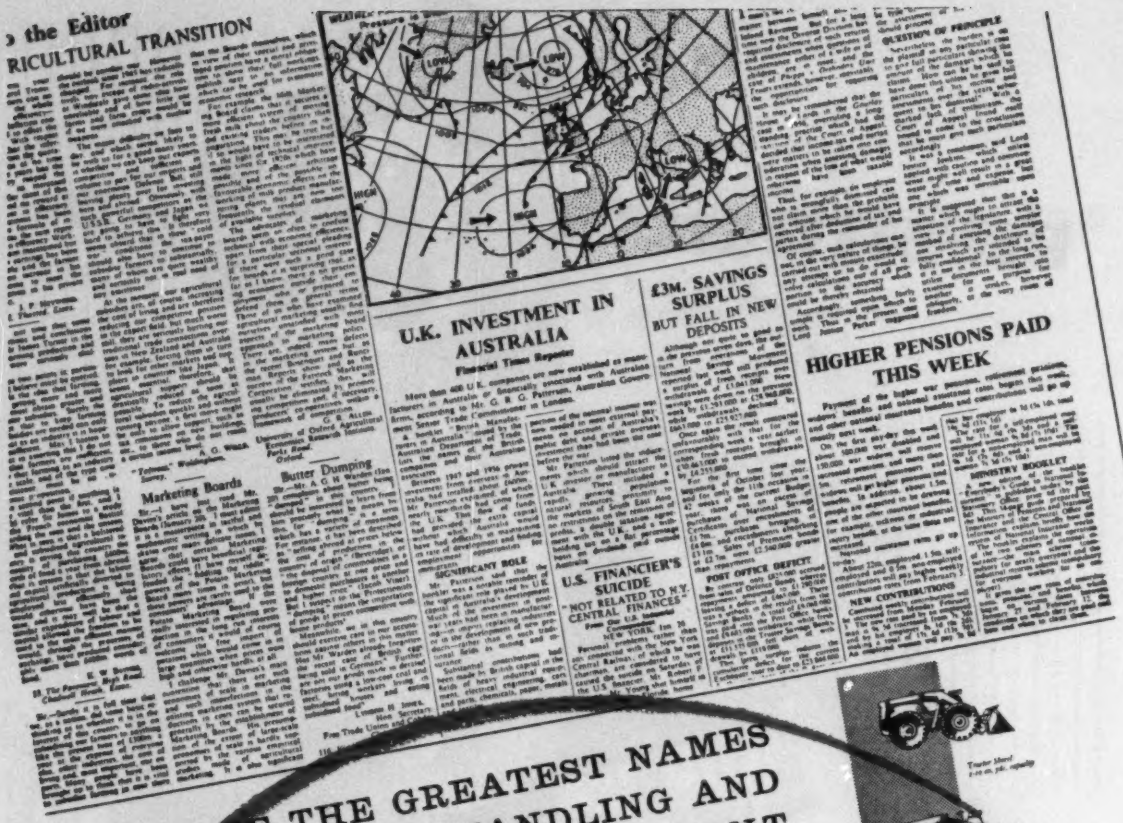

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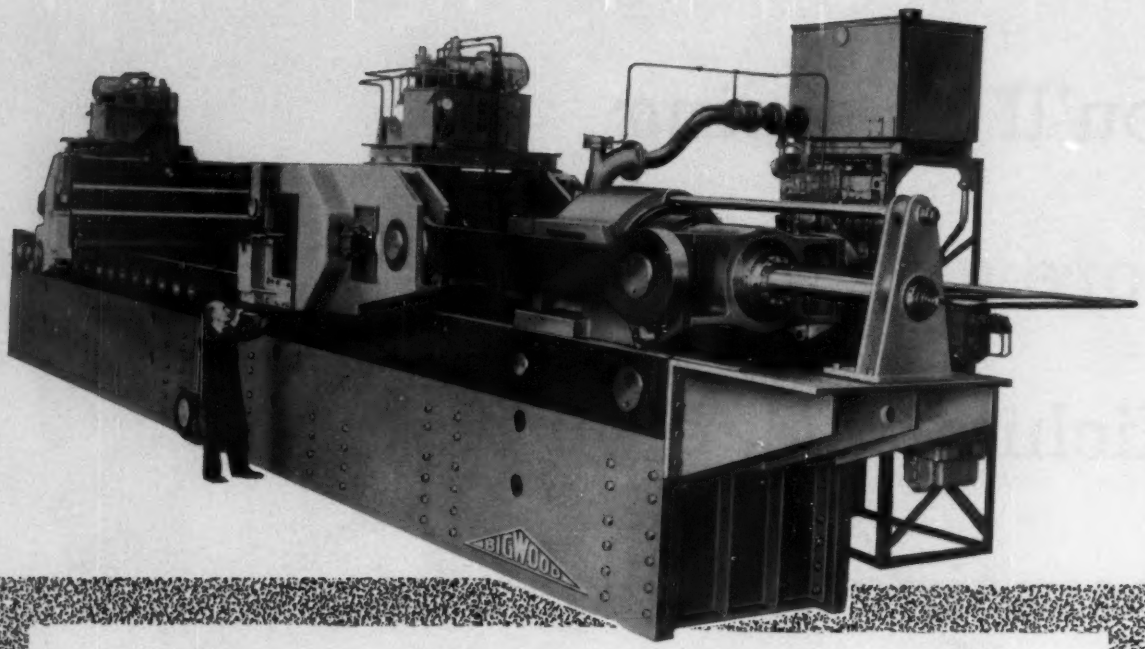

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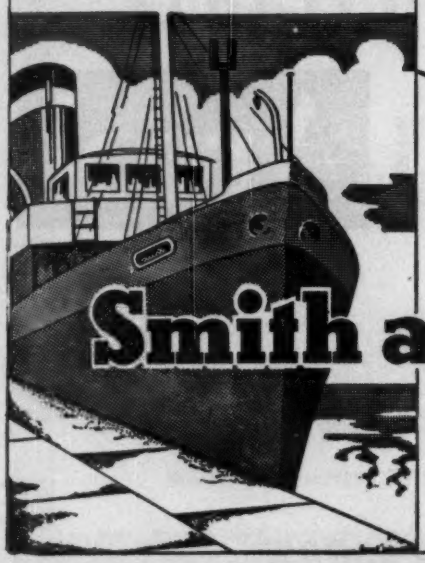

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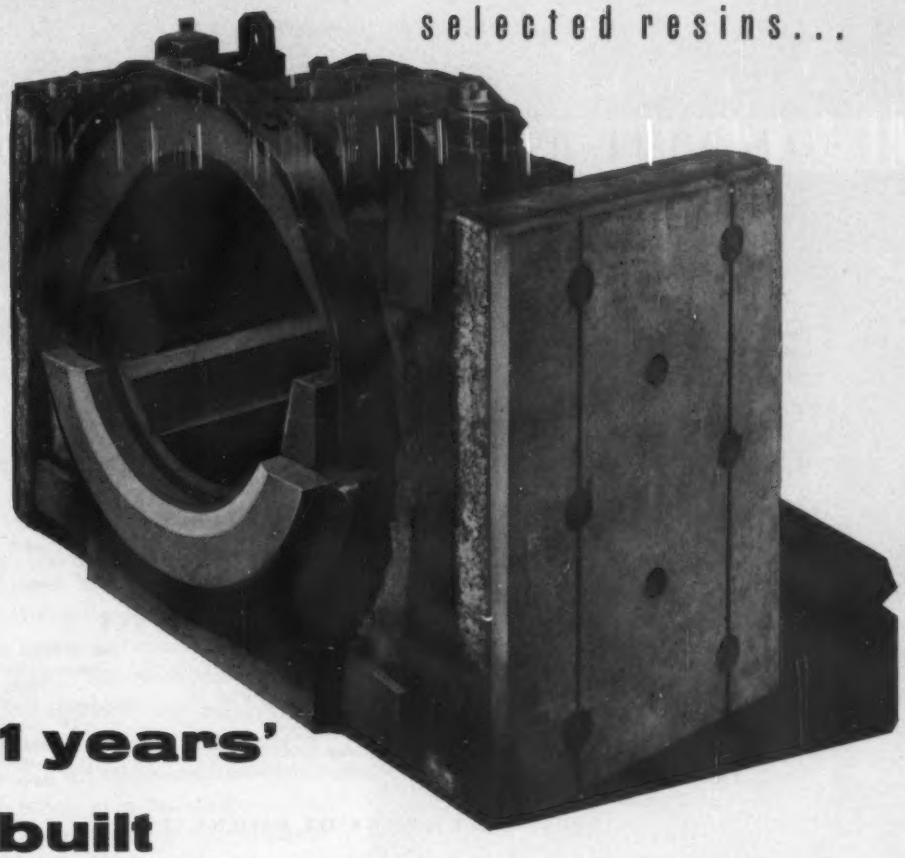

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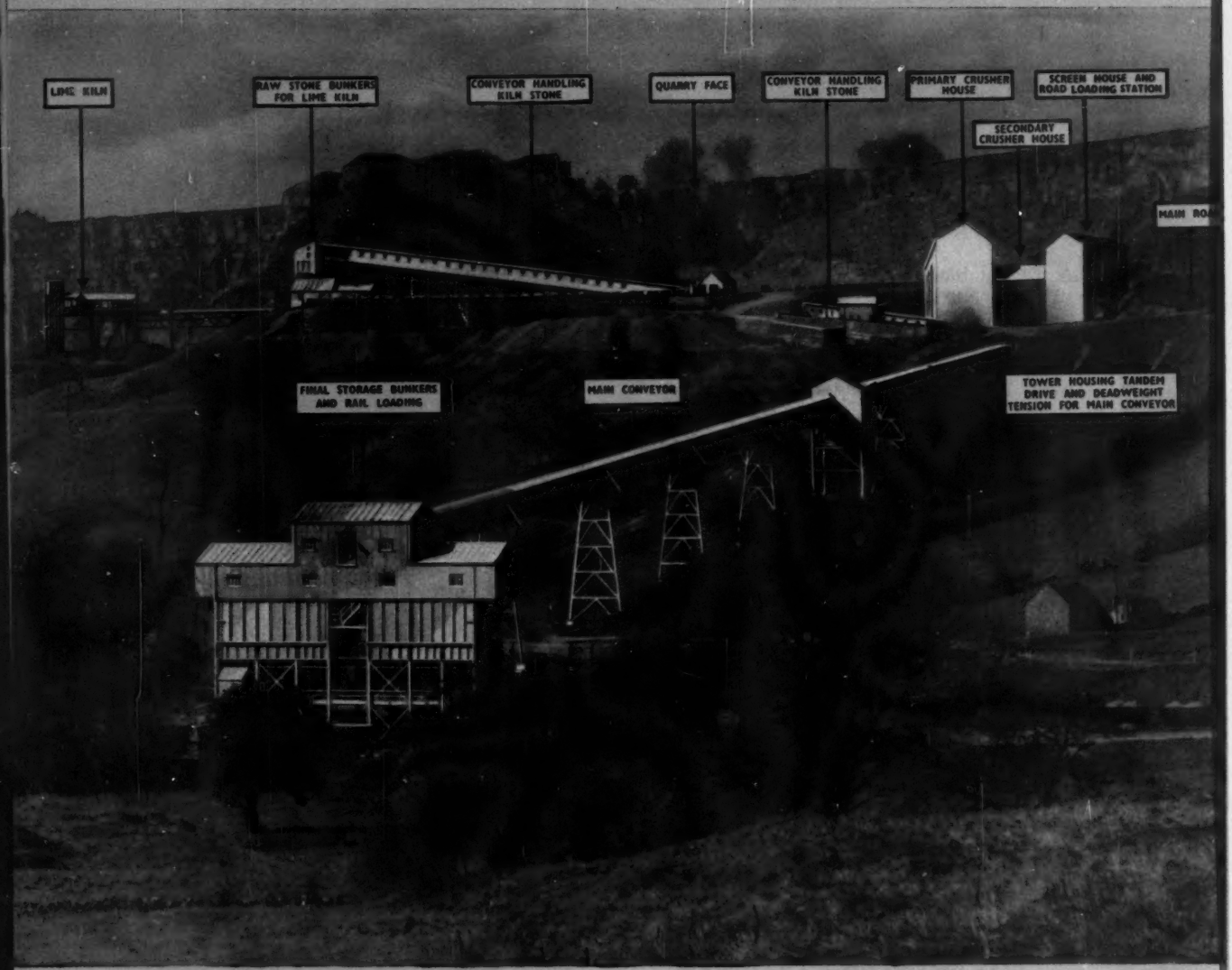
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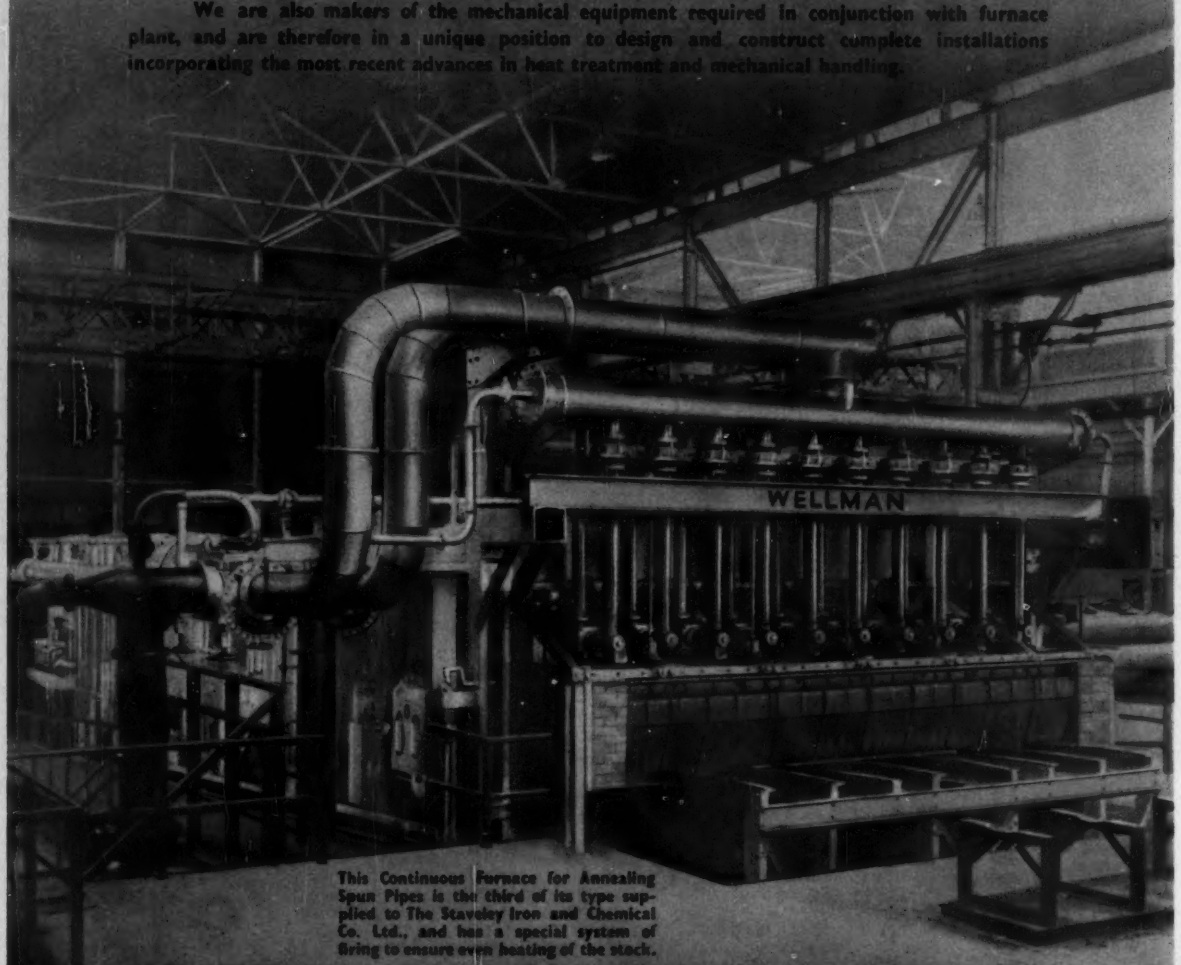

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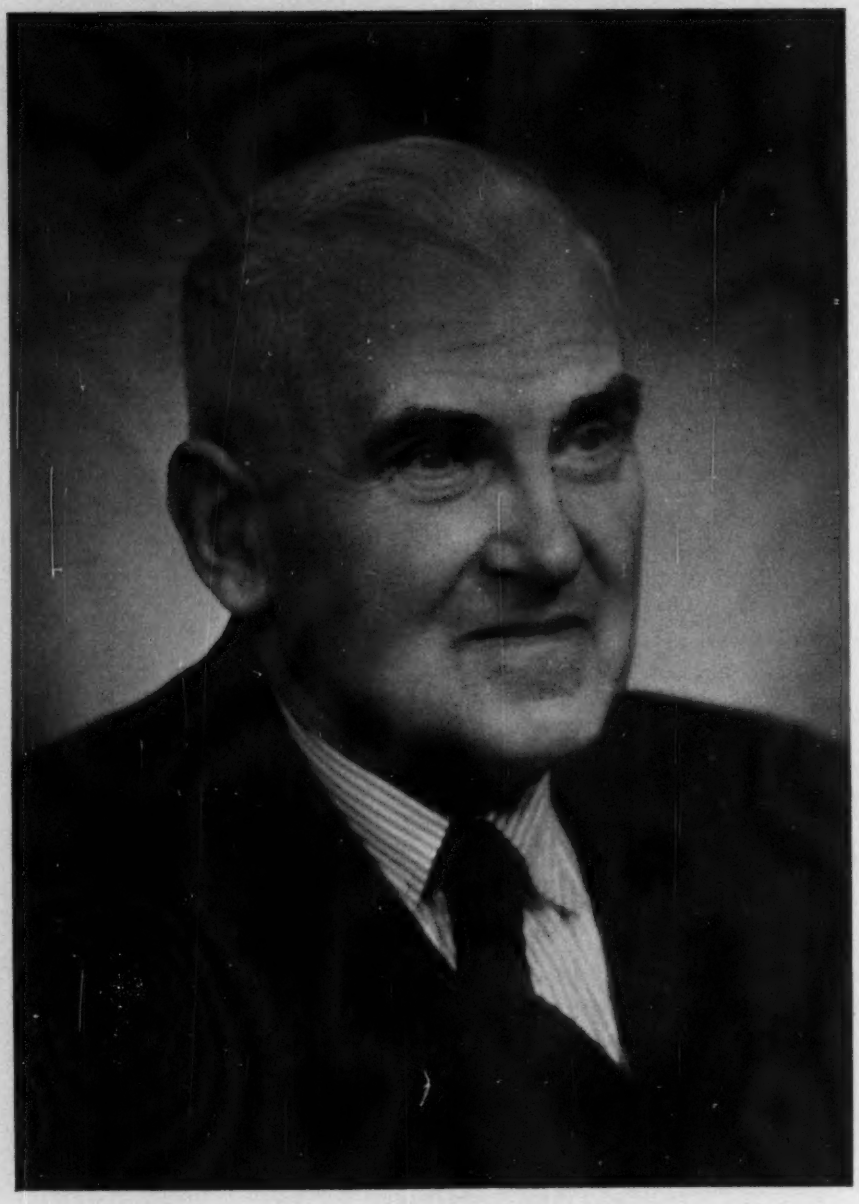
H. R. SCHUBERT, M.A., Ph.D.

DR. H. R. SCHUBERT was born in 1884 at Milisch, in Silesia, the son of a Lutheran clergyman of Swiss ancestry. His mother was a member of the ancient family of Rocholl van Berchem, landowners in the Netherlands and Westphalia. He studied history, economics, and philosophy at various Continental universities from 1903 to 1907, obtaining the degrees of M.A. and Ph.D. He was Assistant at a seminary of the University of Marburg for two years.

In 1909 Dr. Schubert passed the examination enabling him to enter the Berlin Archive Service, and by 1931 he had risen to the rank of State Archive Director. He was deprived of this post in 1933 because of his resistance to Nazism. He lived for some time in Italy and later settled in England with his English wife, subsequently acquiring British nationality.

From 1940 to 1947 Dr. Schubert lectured at the University of Reading. At the end of 1947 he was appointed Historical Investigator to The Iron and Steel Institute. The fruits of this work were a number of Historical Notes, published in the *Journal*, and his " History of the British Iron and Steel Industry from *c.* 450 B.C. to A.D. 1775 ", which was published last year and is reviewed in this issue. Dr. Schubert, who has been a Member of the Newcomen Society since 1948, was invited by the American Iron and Steel Institute in 1952 to act as an adviser for the re-erection of the first American ironworks founded by an English company.

Dr. Schubert's first book, on the palaeography of medieval documents in Belgium and Holland, was published in 1908. Since then he has written many books, articles, and essays on various aspects of European history. His attention was drawn to the history of iron and steelmaking by Dr. Ludwig Beck, himself the author of an extensive " Geschichte des Eisens ", and as a result he wrote two books on the history of the iron industry in Hesse, Nassau, and Siegen. He has recently contributed two chapters on iron and steel to parts IV and V of " A History of Technology " (ed. Singer and others).



H. R. Schubert, M.A., Ph.D.

Historical Investigator to The Iron and Steel Institute

Iron and Steel in Britain

A Review

FROM PREHISTORIC TIMES TO THE INDUSTRIAL REVOLUTION

By Charles Singer

Professor Emeritus Charles Singer, D.LITT., M.D., HON. D.SC. (OXON.), is the author of many works on the history of science and technology. He is Joint Editor of "A History of Technology", currently being published by the Clarendon Press.

1535

DR. H. R. SCHUBERT, in his valuable and interesting work,* distinguishes two periods in the British iron industry. In the earlier, from the beginning until approximately 1500, the 'direct' process only was used. Malleable iron with 0·1% carbon was produced in masses or 'blooms' which had been reduced to paste in a hearth-like furnace but had never become liquid. These blooms of crude wrought iron were hammered together, freed from slag, shaped, and sometimes more or less completely converted into steel. In the second period, the indirect process came gradually to prevail. The ore was treated not in a bloomery but in a much taller 'blast-furnace', fed at the open top and giving a much higher temperature. The liquid metal, combined with 2·2–5% carbon, flowed out as cast or pig iron. This was not useful in itself but had to be 'refined' in a 'finery,' being reduced or smelted down before the iron could be shaped or worked by the forge hammer.

To give his account historical perspective, Dr. Schubert has had to extend his narrative far beyond Britain. In admirably condensed fashion he describes how, from *c.* 1000 B.C., iron had been made in the Austrian Alps. Its classical site there, Hallstatt, has yielded many tools, utensils, and arms. By 450 B.C. Celtic immigrants, bringing the Hallstatt culture, were reaching England from the Low Countries. They settled S.E. of a line from Dorset to Lincoln. Later Celtic settlers between about 250 and about 75 B.C. came from rather further west and brought the more advanced iron culture known as La Tène from a site on Lake Neuchâtel in Switzerland. They penetrated further north than their predecessors and are remembered for their two-horse iron-tyred chariots. The Romans knew well these showy but inefficient military devices and coped with them easily.

From about 75 B.C. Britain was invaded by a third Celtic wave, the Belgae, whom Caesar met in battle in Gaul. They soon occupied Kent, Herts, Essex, and later Hants, Wilts, Berks, and W. Sussex, forming considerable kingdoms, the largest of which came under Cunobelinus, Shakespeare's Cymbeline. They produced elaborate ornamental works, the best known of which are firedogs.

The iron industry has changed the face of Europe in general and Britain in particular. The only fuel for smelting was wood or charcoal. Forest-clad Europe became in time relatively treeless. Deforestation, however, was not merely wastage for, as the trees came down, agriculture advanced.

The composition of early British iron shows that the ores were carefully selected. They were chiefly brown hematite or limonite, mined before Roman times in Somerset and the Forest of Dean in Gloucestershire, red hematite, the best of all, found chiefly in the North West but perhaps not mined till after Roman times, brown siliceous ironstone in West Sussex, and the commoner clay ironstone, while in the north bog iron was often used. All these were long worked as open cast.

The smelting furnace, usual till the Middle Ages in Britain, was a shallow bowl-shaped pit lined with clay, hardened by pounding and by burning. The extraction of metal proceeded in two stages. The first reduced the ore to a spongy metallic state, the bloom. The second consolidated the metal by forging. For the first stage, the ore, mixed with charcoal, was heaped on the hearth and lit. The heat was raised by a blast conveyed by bellows through a clay tuyere. The second stage began when the forgemaster thought the ore adequately reduced. The bloom was extracted with iron tongs and immediately sprinkled with water to facilitate removal of slag by beating with a sledge hammer. The bloom was reheated and then hammered. This was less laborious than has been thought, for the blooms were very small. The reheating, which required temperatures of 850–900° C, was carried on in the same bowl furnace and the blooms were hammered together to form a mass of the desired size.

* "History of the British Iron and Steel Industry from *c.* 450 B.C. to A.D. 1775," by H. R. Schubert. Foreword by the President of The Iron and Steel Institute. 1957, London, Routledge and Kegan Paul Ltd. pp. xxi + 445, with 36 text figs., 11 maps, 26 half-tone figs., and coloured frontispiece. Price 60*s.* net (50*s.* to Members of The Iron and Steel Institute).

Fig. 1—Anglo-Saxon smith at work. (From a ms. of the 11th century, in the British Museum)

Critical for the economics of war and peace was the advent of steel. Its direct production from suitable ores, containing manganese and comparatively free from phosphorus, arsenic, and sulphur, was discovered *c.* 500 B.C. by Celtic tribes in Styria, which became the Roman province of Noricum in the first century B.C. The hold of the Romans on this area partly accounts for the perfection of their arms, a fact which they well appreciated. A small quantity of steel was perhaps produced in Britain by this direct method from the very limited sources of suitable ores. The vast mass of British steel, however, was made by 'cementation,' that is, by wrought iron being converted into steel by absorption of carbon through contact with charcoal at red heat. The depth of carburization varied with the degree of contact and with the duration and heat of the operation. To give suitable hardness to the steel the ancient process of 'quenching' it in water was applied. This also increased brittleness, which was reduced by further heat-treatment or 'tempering.' Steel was expensive and therefore economized. A device to this end, widely applied until late medieval times, was 'steeling,' that is, welding a strip of steel on a wrought-iron base.

The Romans, as Dr. Schubert shows, diffused the iron industry in Britain over a wide area; they introduced true mining, improved the steel product, and facilitated smelting by raising the height of the furnace. This made the insertion of a tuyere and manipulation of the furnace simpler and more effective. In making their roads, the Romans found a use for slag and, at some spots, the slag has so concreted that the surface is still intact. This use of a waste product was important, as Dr. Schubert well brings out. A Roman novelty was a special second furnace for reheating and welding of blooms into large masses. This is seen in a works of *c.* 300 A.D. at Corbridge-on-Tyne near the Roman Wall. Cast iron, if produced, was an accident.

Of the Anglo-Saxon iron industry we know little, though English cutlery gained a continental reputation. There are beautiful examples of Anglo-Saxon swords with decoration and runic lettering inlaid with brass or silver of the ninth to the eleventh centuries. Important outlets for smith's work were coulters and shares for ploughs. Smiths often appear in legend and occasionally in the art of late Saxon mss. (Fig. 1). No remains are known of furnaces from the period, but cementation and steeling continued. Both seem to have been improved by the Danes.

The industry suffered a setback in the Norman period (1066–1154), because of the wastage of land and men in the wars before and after the Conquest. Most surviving works passed into the hands of the nobles. From this depression the industry was lifted by the Cistercian monks whose order was founded in 1098 at Cîteaux in Burgundy. In 1128 they settled in Surrey. Soon monasteries of the same order were formed in the North, notably at Furness in Lancashire and Rievaulx and Fountains in Yorkshire. The Cistercians were great mechanics and continued active during the disturbed reign of Stephen (1135–1154), acquiring mines and erecting ironworks wherever ore was available. In the north they contributed to the systematic exploitation of clay ironstone. But their main contribution was to the solution of the problem of power by the introduction of water-driven wheels. The technological key to the economic situation has always been *power*. In accord with their history, this development was almost certainly due to the Cistercians of whose work Dr. Schubert gives many extremely interesting facts.

Towns gained in importance from the twelfth century onward. This meant increased demand for iron, especially for building. The chain armour of the period called for special skill from the smith and an exceptional quality of metal. The most elaborate ironwork produced by the Anglo-Norman smith were the grilles which encased tombs or enclosed choirs and side chapels. Dr. Schubert has figures of very beautiful examples of these grilles.

Between 1154 and 1348 there was a rise and then a decline in the iron industry. The increase of population, especially in the towns, meant more building. There was an expansion of overseas trade with the rise in the export of wool, the great British staple of the time. The Crusades, together with the improvement in agriculture, also meant a rise in the demand for iron and steel. The Forest of Dean and the adjoining Wye valley developed rapidly as mining and ironmaking centres. This activity extended well into Wales, Somerset, and Wilts. Moreover, the country between Sheffield and Leeds became a great iron-yielding area in the thirteenth century. Further north, in Lancashire, a Cistercian house, the Abbey of Furness, led the industry from 1235. Thus many forges

arose, though each had only a very small output of the order of 3 tons a year. This was inadequate to the needs of the time so that there developed a considerable import from Styria, Normandy, Spain, and Sweden. From the last came especially the very high quality of iron known as 'Osmund'—a word of unknown origin but familiar to the medieval merchant as an indication of value.

A demand for steel was steadily increasing from the mid-thirteenth century, especially for ploughshares, ploughshoes, the axles of carts, parts of horseshoes, for the machines of mills, now rising all over the country, as well as for marine needs such as anchors and chains. Moreover town crafts needed steel tools with sharp edges and points for building and carpentry as well as for arms and armour. Arrows were now pointed with steel.

A decline in the industry set in about 1300. Northern activity was checked by repeated raids of the Scots, the woodland resources were everywhere shrinking, and the number of forges was already falling when there came the disaster of the Black Death of 1348. The shortage of labour that followed put a premium on mechanical devices to replace human power. It is probably for this reason that the waterwheel for working furnace bellows and hammers became familiar in England about this time. By 1400 there were waterwheels in many English bloomeries. At the same period we hear too of mechanical tilt hammers raised by the pressure of the foot and then let drop and mechanical devices for regulating smith's bellows (Fig. 2).

Towards the end of the fifteenth century we have news from Britain of the indirect method of iron and steel-making and its characteristic implement, the blast-furnace. This was of considerable height and was fed at the open top. The liquid metal which it continuously produced was tapped from time to time as required. The resulting pigs of cast iron had to be smelted down again in a 'finery,' a new intermediate stage.

Between about 1480 and 1580 we have from continental sources a number of representations of the iron and steel industry, some of which are attractively set out by Dr.

Fig. 2—Blacksmith's forge with bellows controlled or worked by a lever in the hand of the smith. (From the Holkham Bible, c. 1330, in the British Museum)

Fig. 3—Bloomsmithy, c. 1550, from Agricola. *A* is the hearth. The smith wears a mask against fumes. He controls the bellows just seen behind with a lever. *C* is the slag vent, *D* is a mass of blooms being hammered, *F* is the power hammer, presumably worked by a waterwheel, and *G* is the anvil.

Schubert (Fig. 3). In 1490 the term 'iron foundere' first occurs in English records. In 1496 Henry VII commissioned one Henry Fyner (doubtless a 'refiner') who had been a goldsmith in London, to engage as many 'founders' and labourers as were needed to erect buildings for making iron for use in war with the Scots. Fyner worked in Ashdown Forest in Sussex. His wrought iron was forged by a 'great water hammer.' He produced sections of wheel rims ('felloes'), pegs for connecting them, bars for axles, wedges etc. Pig was produced in a blast-furnace and after refining was delivered to the King's gunner at the Tower, who used it for casting gunshot of from 5 to 224 pounds.

The first half of the sixteenth century saw a great development of ironworks and steel forges in the Weald of Sussex, notably around Robertsbridge. Many of the workers were foreigners, French or Flemings. From this time begin to appear the cast-iron firebacks and tombstones, familiar as museum-pieces. Many are illustrated by Dr. Schubert.

An important use of iron from the fourteenth century onward was in the making of guns and shot. By the

Fig. 4—Blast-furnace with blast worked by undershot waterwheel and enclosed in a wooden shed. Second half of eighteenth century. (From "La Grande Encyclopédie" of Diderot)

end of that century cannons were being made from wrought-iron bars welded together on a core. The largest were of about 600 pounds. About 1535, Henry VIII, fearing invasion by the French, emplaced many guns on the south coast. They were of cast iron, which was less expensive than wrought iron. To make these he employed a Flemish director (1534–1539), who in 1541 was succeeded by the Rev. William Levett, Rector of Baxted in Sussex, who held the post till his death in 1551. It seems an odd position for a parson, but he was experienced in the trade and became a highly efficient administrator. Dr. Schubert gives a very entertaining account of Levett's extra-clerical activities. His main works were in his own parish and employed many French workmen. He was succeeded by his French manager who continued active into the eighties. The making of cannon initiated a boom in the Weald where by 1548 there were 20 blast-furnaces and 28 forges. By 1550, Sussex works were actively exporting but rapid deforestation soon brought a decline. By 1560 the indirect process was spreading to the Forest of Dean and neighbouring areas and to the Midlands. By the end of the century it had reached the North.

The first half of the seventeenth century saw extensive development of the industry in Britain. The bloomeries had now been generally replaced by blast-furnaces. These demanded more capital and the great iron masters began to appear. Two members of this class, of very different types and fates, were William Herbert, Third Earl of Pembroke (1580–1630) and Richard Boyle, First Earl of Cork (1566–1643). Pembroke had vast estates in South Wales and Monmouth and interests in the Forest of Dean, of which he was warden. He erected many blast-furnaces and forges in the forest under a Royal grant of 1612 and cut down prodigious quantities of timber. In a few years, however, he made over his leases to tenants who worked on a smaller, but not more economic scale. Despite extensive building, the works were in a poor state by 1635 and the Crown intervened. To save fuel the number of works was restricted and the bloomeries, which still survived in the Forest, dismantled. In 1640 all leases were surrendered to a nominee of the secretary to the wife of Charles I. The Civil War was now approaching. In 1644, the lessee, naturally an ardent Royalist, was attacked by the Parliamentary troops and the whole ironworks controlled by him were destroyed.

Richard Boyle was a most able and efficient go-getter. With a buccaneer's morals, which mellowed with advancing years, he sought his fortune in Ireland. There he found and lost more than one but became immensely wealthy, acquired the title of Earl of Cork, and developed in some respects into a model landlord and captain of industry. Of his 15 children, the fourteenth was the saintly Robert, "father of chemistry." Richard Boyle's rise was fortunate for Ireland for he was a very great employer of labour, setting up numerous works. His example was followed by several others. Cromwell, when he saw the prodigious improvements that Boyle had effected, said that had there been one like him in each province, the Irish Rebellion would have been impossible.

During the second half of the seventeenth century, the blast-furnace was considerably improved and modified in form (Fig. 4). It was usually enclosed in a shed largely of wood, despite danger of fire. The new type stood some 20 ft or more above ground and its inner casing was round instead of square. Its 'tunnel' widened downward to the 'boshes,' where it narrowed again to the 'hearth' where the tuyere was inserted. The narrowing of the top prevented escape of heat. The construction of the hearth was important and a special care of the founder. Wooden box bellows replaced the old type. They consisted of two closely fitting boxes, one of which was raised or lowered on the other. They were more durable and cheaper to work than the old type.

A seventeenth century improvement was the wider use of fluxes to promote slag formation. Among the fluxes used were limestone and slags or cinders from old works. Slag was exported to Ireland, mainly to the estates of the Earl of Cork.

Smelting was described by several eighteenth century writers. The spectacles that most attracted them were the torrent of liquid iron that gushed forth from time to time from the bottom of the furnace and the flames from the aperture at the top. Interest in the subject was focused by the classical and beautifully illustrated *Memoirs on Steel and Iron* (1722) by Réaumur (1683–1757). Even more remarkable were the *Opera philosophica et mineralia* (1734) of the mystic, Emanuel Swedenborg (1686–1772), whose feet were very much on the ground in all that concerned his profession of mining engineer. He spent many years in England. The forms of furnaces used in this country are described and figured in the *Commercium philosophico-technicum* (1763) of William Lewis (1741–81) and in a manuscript by that author which is worth publishing in full, though a beginning has been made by Dr. F. W. Gibbs.

This brings us to the end of the period treated by Dr. Schubert. It is, however, by no means the end of the subjects that he discusses. We have said almost nothing of the processes of casting and boring, of his account of the construction of the furnaces and other buildings or of the chemistry of the processes, of his wealth of statistics, or of the artistic aspects of the subject and of the works of art illustrating it to which he draws attention. He has made a most important contribution to the all too scanty literature of the technology, the heritage of which we are beginning to reap. We end on our only critical note. It is that we hope that future editions will be provided with better maps. It is not sufficiently appreciated that the making of maps is a professional task and their adequate presentation helps and adds to the pleasure of the reader as well as to the beauty of the volume.

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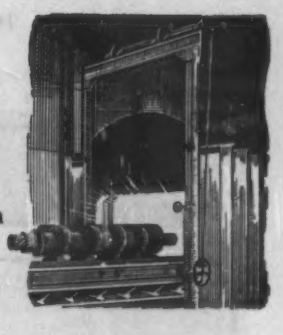


The Closing of Internal Cavities in Forgings by Upsetting

By A. Tomlinson, A.Met., A.I.M., and J. D. Stringer, B.Sc.(Tech.)

SYNOPSIS

Upsetting as a means of closing internal cavities in forgings has been investigated using small-scale models containing artificial cavities. The effect of upsetting on cavities has hitherto been uncertain, but the results of the present work suggest that all cavities could be closed by upsetting, following the method described in the paper.

Upsetting tools with dished working faces have been developed which would enable upset pieces to be produced not only free from cavities, but also of a suitable shape for reforging. The tests suggest that the reforged pieces would have satisfactory mechanical properties at the positions of healed cavities. 1481

UPSET FORGING consists of reducing the height of a forging and increasing its diameter. It may be used on dimensional grounds (i.e. for producing forgings of large diameter relative to the initial ingot diameter), or for improving the mechanical properties of forgings in the transverse directions.

The effect of upsetting on shrinkage cavities present along the axis of a forging ingot has not previously been understood, and on the recommendation of the Forging Committee of B.I.S.R.A. the value of upsetting in this respect was investigated on a model scale. This paper describes the effects of upsetting mild-steel blocks containing machined holes in positions where shrinkage cavities normally occur. Most holes were situated along the axis of the test blocks, but some were slightly displaced from the axis.

The application of the results to practical forging operations is discussed and a suitable procedure for upsetting and reforging is given.

EXPERIMENTS

Upsetting

The forging tools used for upsetting completely overlapped the ends of the test blocks. Normal upsetting procedure was followed. All tests were done at 1200–1250° C on a 200-ton press, each operation being completed in a single squeeze using a nominal penetration speed of 1½ in./s.

Flat Tools

For the initial tests, 4 in. square blocks in 4-in. and 8-in. lengths were used. A ¼-in. dia. axial hole was drilled through each block to within ¼-in. of one end, and the open end was plugged. The blocks were then upset to various degrees. On sectioning the blocks it was seen that the holes closed up initially at the mid-height of the block and that as the degree of upset was increased, so closure extended towards the ends (*see* Fig. 1). The longer blocks required more upsetting to close the holes than the shorter ones. In the longer blocks the holes began to close after the height was reduced to 0·36 of the original, whereas in the shorter blocks closure began when the height was 0·58 of the original. It was noted at this stage that in both sizes of block the width (diameter) of the upset piece in contact with the tools was almost exactly twice the height. Further tests showed that closure was virtually complete, in both sizes of block, when the height/diameter ratio reached 1 : 4.

Further tests with smaller axial holes of $\frac{1}{16}$, $\frac{1}{8}$, and $\frac{1}{16}$ in. dia. gave similar results, except that the holes with smaller diameters closed a little sooner; for example, the $\frac{1}{16}$-in. holes began to close when the height/diameter ratio was 2 : 3, instead of 1 : 2 as for ¼-in. holes. Other tests showed that holes slightly displaced from the axis closed at similar height/diameter ratios, even though they bent outwards at the mid-height. The results shown in Fig. 2 are from holes originally ¼-in. away from the axis. Displacements greater than this are not of practical interest.

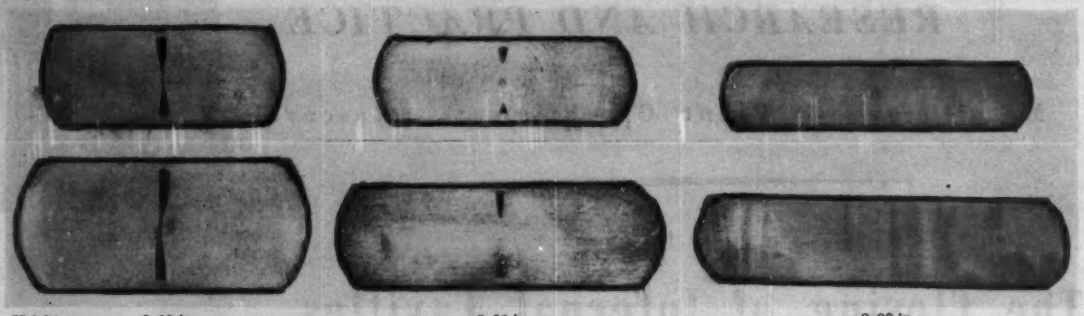
Next, 4 in. dia. × 8 in. long blocks, each with a tapered extension at one end to represent an ingot

Paper MW/F/36/57 of the Forging Committee of the Mechanical Working Division of the British Iron and Steel Research Association, received on 1st October, 1957. The views expressed are the authors' and are not necessarily endorsed by the Committee as a body.

The authors are on the staff of the Mechanical Working Division at the Association's Sheffield Laboratories.

Blocks initially 4in. long

Height	2·28 in.	2·00 in.	1·5 in.
H/D ratio	0·51	0·41	0·25



Height	3·13 in.	2·50 in.	2·00 in.
H/D ratio	0·60	0·40	0·27

Blocks initially 8 in. long

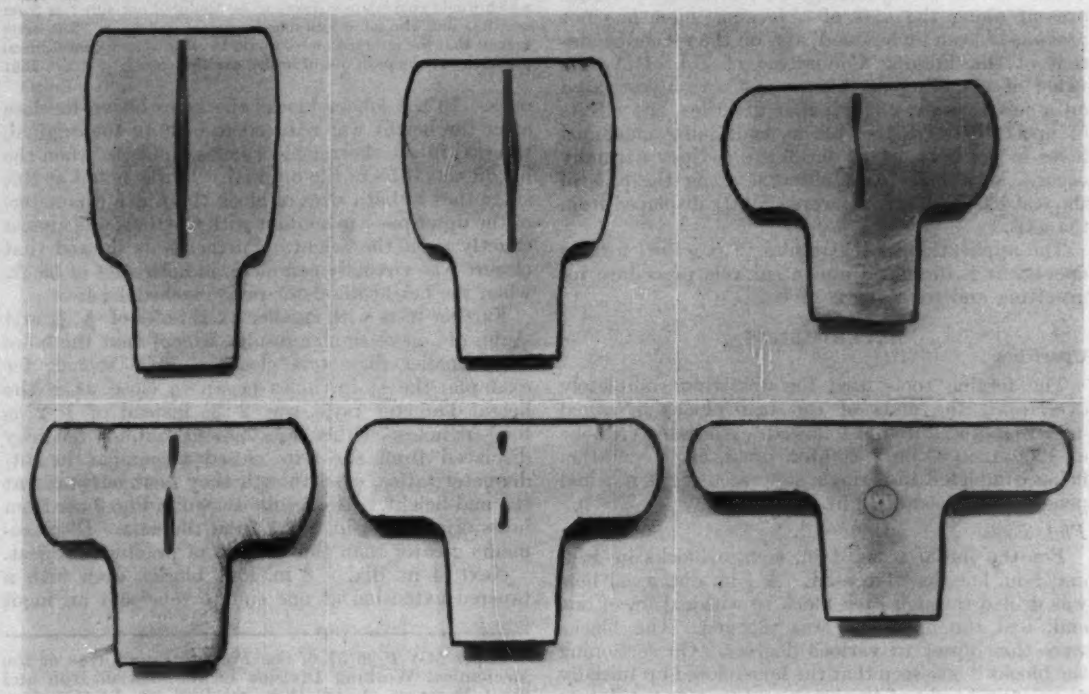
Fig. 1—Tests on blocks containing a ¼-dia. axial hole

Blocks initially 8 in. long

Height	3·19 in.	2·47 in.	2·00 in.
H/D ratio	0·65	0·40	0·27

Fig. 2—Tests on blocks containing a ¼-in. dia. " offset " hole

Body height	5·9 in.	4·84 in.	3·1 in.
H/D ratio	1·44	1·12	0·57



Body height	2·88 in.	2·42 in.	1·7 in.
H/D ratio	0·50	0·39	0·21
% closure	Healing just commenced	47	100

Fig. 3—8-in. long test ingots upset between flat tools

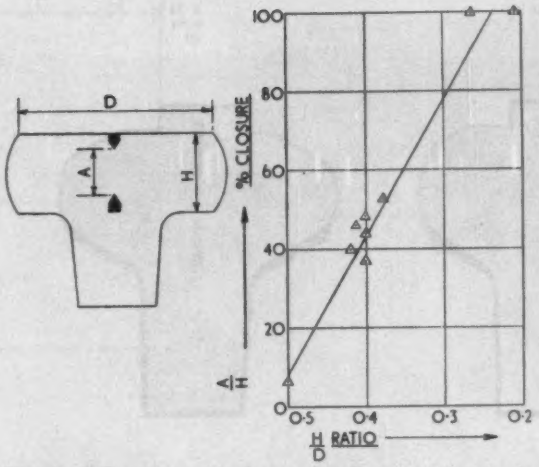


Fig. 4—*H*/*D* ratio *v.* % closure for flat upsetting tools

head, were upset. The blocks initially contained a $\frac{1}{4}$-in. dia. axial hole, and the effects of upsetting on the holes are illustrated in Fig. 3. Evidently the extension piece did not affect the closure obtained. In these experiments it was noted that at first the hole became enlarged in its central portion while the block was still longer than its diameter; no enlargement was observed in blocks with an initial height/diameter ratio equal to or less than 1. By taking measurements throughout the series of tests, the length of hole closed was correlated with the height/diameter ratio, as shown in Fig. 4.

Measurement of the forging loads required during the above tests showed that the production of a piece in which the axial hole has been completely healed by upsetting required the full capacity of the laboratory press, i.e. 200 tons.

Dished Tools

The pieces upset between flat tools were of an unsuitable shape for reforging because of the risk of overlap at the ends; to obviate this defect, the then Chairman of the Forging Committee (Mr. R. C. Benson of The Darlington Forge Ltd.) recommended the use of dish-shaped tools for upsetting. It was anticipated that the shapes produced in these tools would be satisfactory for reforging. Figure 5 shows a drawing of the experimental tools with a test block assembled ready for upsetting. The circular working faces were at an angle of 30° to the horizontal. The dimensions, which depend on the volume of smallest ingot to be used, were calculated using standard formulæ for the volumes of conical shapes.

Two series of ingot models, with bodies measuring 4 in. dia. and 8 and 10 in. long respectively, each containing a $\frac{1}{4}$-in. dia. axial hole, were upset to various degrees in these tools. The load needed for each operation was measured. As shown in Fig. 6, upsetting in the dished tools caused the holes to close at an earlier stage than in flat tools. The load required to close the hole completely was 200 tons. Similar results were also obtained with ingots models having a 30° tapered base and a larger radius at the junction of the head and body.

As with flat tools, the proportion of the hole closed depended only on the final shape after upsetting. This was ascertained as follows. The degree of closure was defined as the ratio of the length of hole which had closed to the current length of the ingot body. The final shape was defined in terms of a height/diameter ratio in which the height was calculated by projecting lines from the working faces of the tools, as shown in Fig. 7, and the diameter by measuring the contact distance between the tools and the upset piece. By so calculating the amount of closure and the height/diameter ratio, it was found that results from ingots with different initial sizes and different shapes at the head and base fitted on to a single curve (Fig. 7).

Short Holes

With a view to simulating real ingot cavities more closely, $\frac{1}{2}$-in. long holes of $\frac{1}{32}$ and $\frac{1}{4}$ in. dia. respectively were introduced at the centre of blocks 4 in. dia. × 8 in. long and the blocks were upset as before. These blocks were prepared without head or base extensions to save expense, but instead cold pieces of metal were inserted in the tool recesses which normally accommodate the head and base of the ingot. Figure 8 illustrates both the mode of construction and the results of upsetting.

At first, the $\frac{1}{32}$-in. dia. holes increased in diameter and shortened in length, but as upsetting was continued they began to shrink in diameter. They finally closed after the same upsetting reduction as was needed to close the central portion of a full-length hole of this diameter. The $\frac{1}{4}$-in. dia. holes also increased in diameter and shortened at first. Subsequently, however, their ends welded together without any sign of shrinking. The welding was completed when the body length was about 4·9 in.;

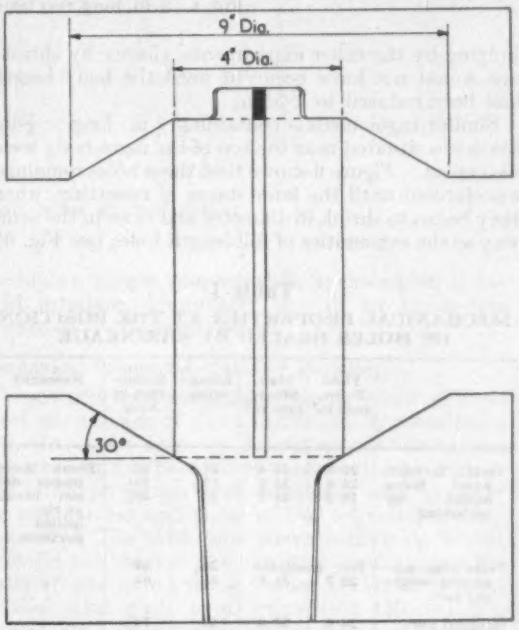


Fig. 5—30° dished upsetting tools

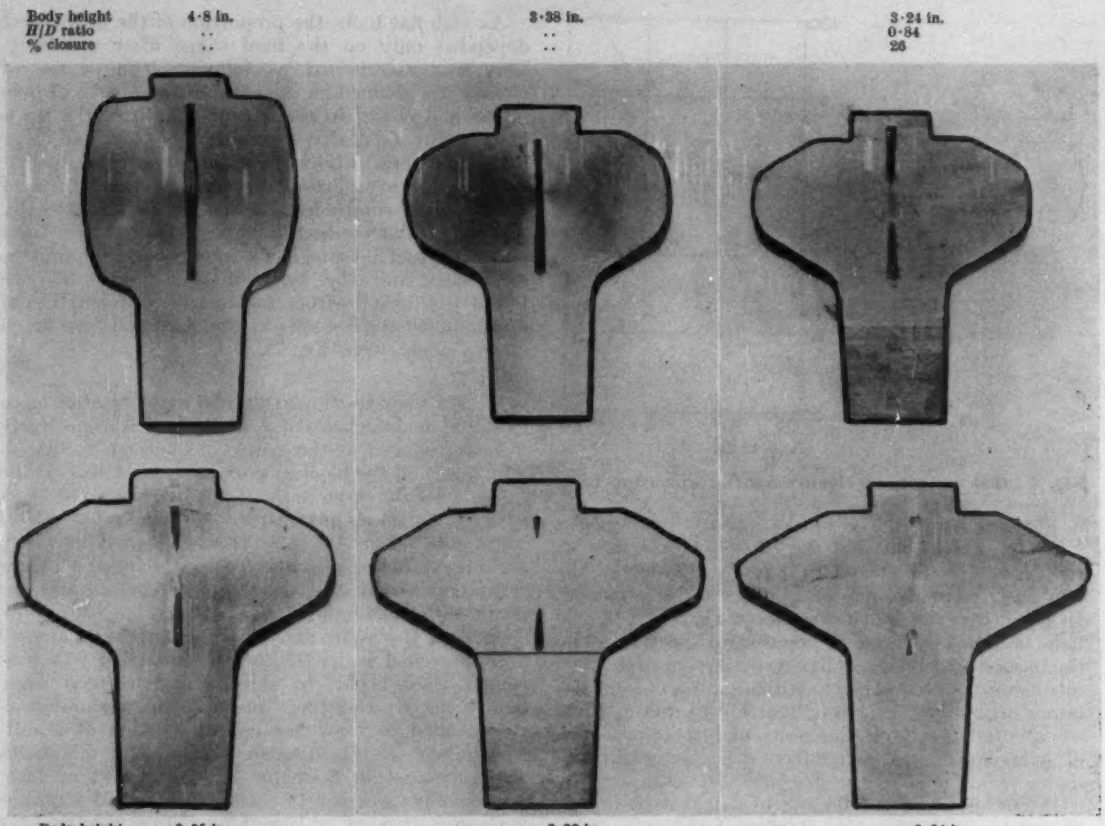


Fig. 6—8-in. long test ingots upset between dished tools

judging by the other experiments, closure by shrinkage would not have occurred until the body length had been reduced to 3·35 in.

Similar ingot models containing $\frac{1}{2}$-in. long × $\frac{1}{4}$-in. dia. holes situated near the top of the ingot body were also upset. Figure 9 shows that these holes remained undeformed until the later stages of upsetting, when they began to shrink in diameter and close in the same way as the extremities of full-length holes (*see* Fig. 6).

Table I

MECHANICAL PROPERTIES AT THE POSITION OF HOLES HEALED BY SHRINKAGE

	Yield Point, tons/in²	Max. Stress, tons/in²	Elongation, %	Reduction in Area, %	Remarks
Tests through axial holes healed by shrinking	22·0 22·0 19·2	36·4 35·5 35·2	28½ 19½ 15	64 39½ 39½	These three pieces did not break at the healed portions
Tests after upsetting original bar	Not available 23·7	 35·9	24½ 31½	49 62	
Original bar	24·0 22·0	35·4 35·5	12 23½	14½ 36½	

Mode of Closure of Holes

All holes extending the full length of the blocks were closed by shrinking in diameter; so were the

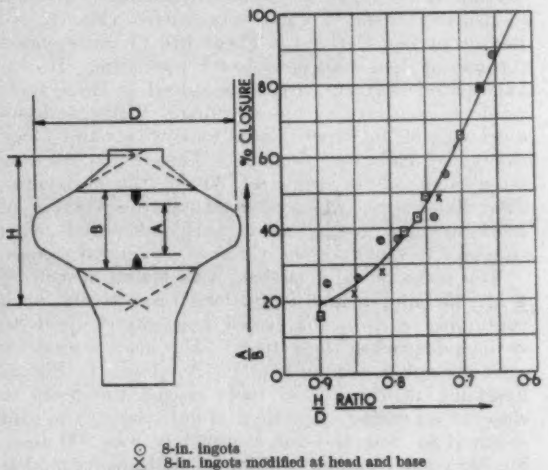


Fig. 7—*H/D* ratio *v.* % closure for dish-shaped upsetting tools

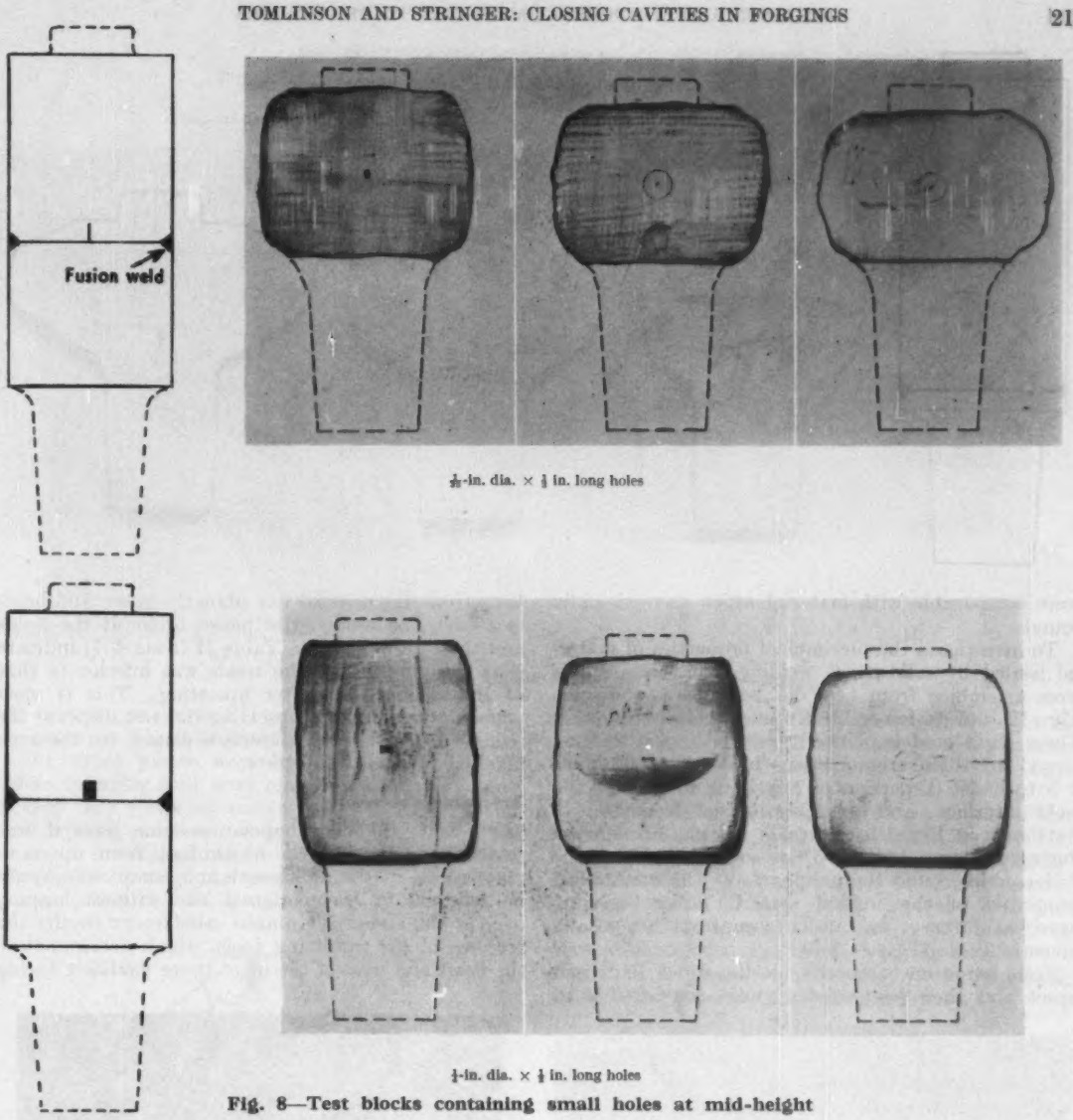


Fig. 8—Test blocks containing small holes at mid-height

central holes measuring $\frac{1}{32}$ in. dia. × $\frac{1}{4}$ in. long. Figure 10 shows a transverse section through a healed $\frac{1}{4}$-in. dia. full-length axial hole after etching in 2% nital. The metal has moved radially inward to the axis of the hole. In the centre of the photograph is a minute filament of scale due to oxygen entrapped in the hole during preparation.

A different effect was obtained with the $\frac{1}{4}$-in. dia. × $\frac{1}{4}$ in. long holes situated at the mid-height of the block. In this case, the holes decreased in length and the end faces met and welded together. The weld interface is difficult to detect with the etching reagents in general use, but Fig. 11 shows the location of a weld (between the scratches) after etching in an alkaline chromate reagent.* This reagent indicates the higher oxygen concentration in the region of the weld interface, presumably due to air which was entrapped during preparation.

Mechanical Properties of Healed Cavities

Transverse tests were taken from the centre of three upset pieces, each of which had originally contained a $\frac{1}{4}$-in. dia. axial hole, to determine the mechanical properties at the position of holes healed by shrinkage. The properties were compared with those of the original bar and those of the original bar after upsetting. The three test pieces containing healed holes did not break in the vicinity of healing, and the material was found to be harder (171–172 D.P.N.) at this point than $\frac{1}{2}$ in. away (152–159 D.P.N.), presumably because more plastic flow had occurred near the hole. As Table I shows, the properties

* L. Fine, *Metal Progress*, 1946, Jan., pp. 108–112.

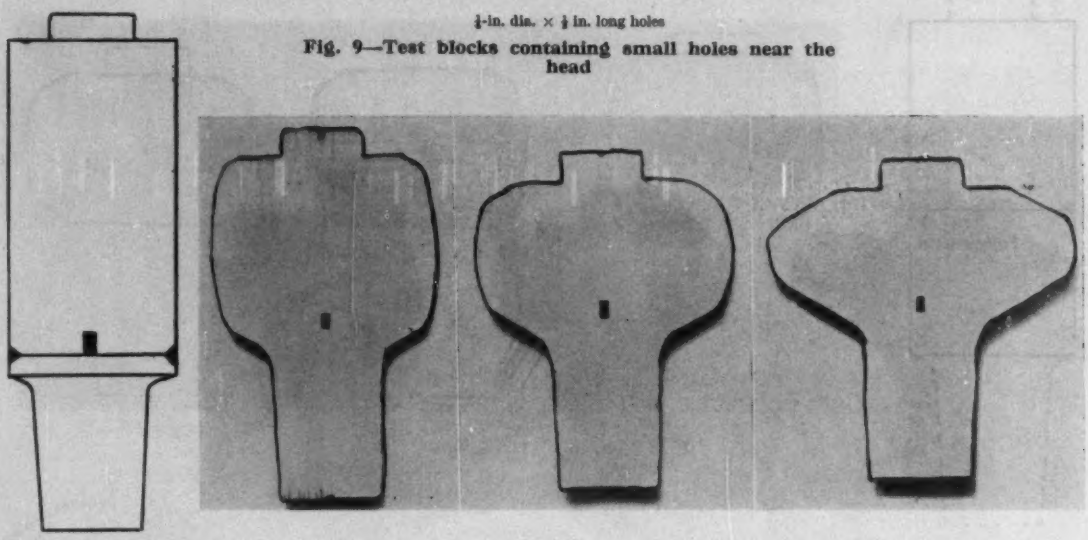


Fig. 9—Test blocks containing small holes near the head

were comparable with material which was originally sound.

To investigate the mechanical properties of material healed by solid-phase welding, composite ingots were assembled from 4-in. dia. sections as shown in Fig. 12 and fusion-welded round their peripheries. Three ingot models of this type were upset and reforged to their original size, held at 1250° C for 2 h to assist dispersion of any scale formed at the weld interface, and then normalized from 900° C. As shown in Fig. 12, the welds did not break open during reforging, but there was some cracking at the fusion welds round the peripheries. The mechanical properties of the welded material after reforging were satisfactory, as can be seen from the results given in Table II (tests 1–3).

Tests made on composite blocks, which had been upset and then heat-treated, were compared with tests from the original bar similarly upset and heat-treated. The welded test pieces broke at the welds and the results given in Table II (tests 4–7) indicate that the ductility of the welds was inferior to that of the original bar after upsetting. This is most probably because the upsetting did not disperse the oxide film at the weld interface caused by the artificial conditions of the tests.

Reforging Procedure

By using the dish-shaped upsetting tools it was established that pieces, which had been upset to close up a ¼-in. dia. full-length hole completely, could be reforged to their original size without lapping over at the ends. To obtain satisfactory results the recesses in the upsetting tools, which accommodate the head and base of the ingot, were modified during

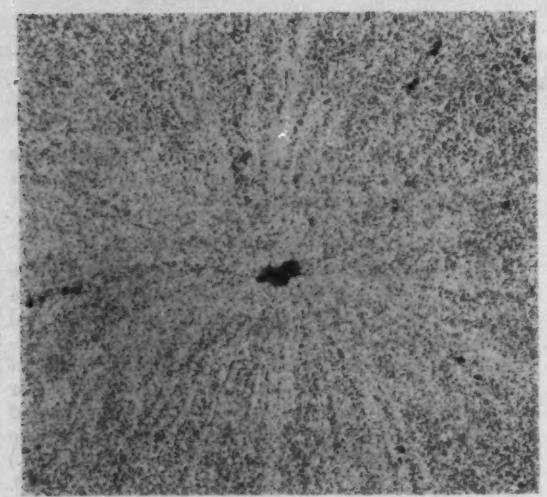


Fig. 10—Healed axial hole after etching (2% nital) × 35

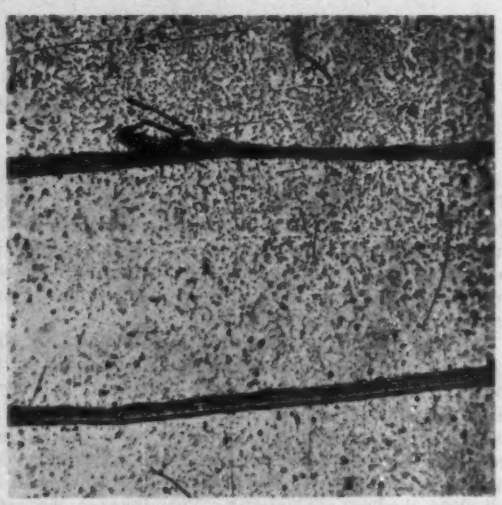


Fig. 11—Solid-phase weld after etching (alkaline chromate) × 35

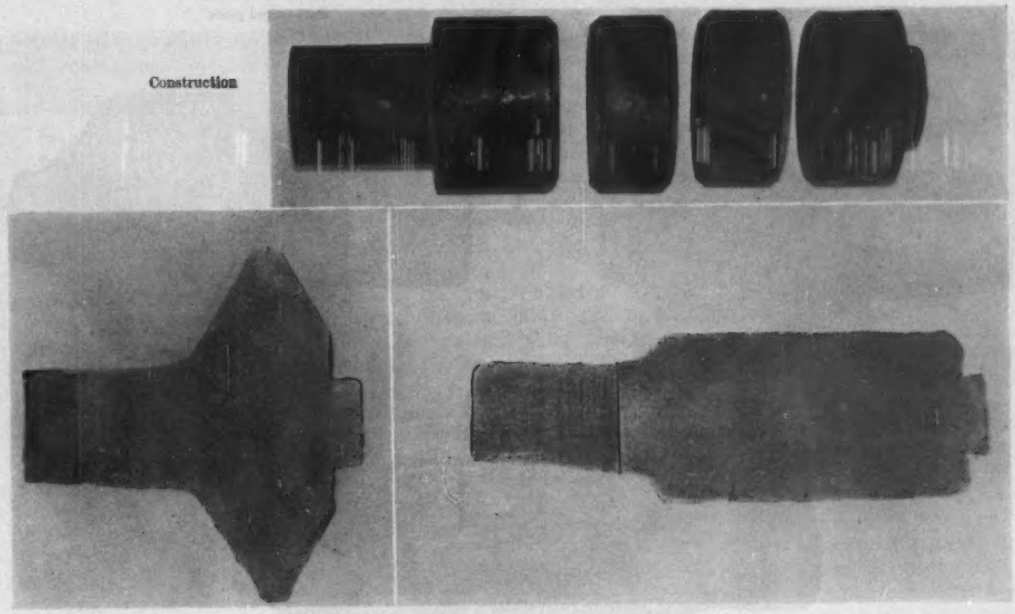


Fig. 12—Composite ingots with solid-phase welds

the reforging tests, and tools with suitable contours are shown in Fig. 13 (compare with Fig. 5).

The upset pieces were heated to 1200–1250° C before reforging and were reheated during reforging to keep their temperature above 850° C. It was found that lapping was avoided by continuing the initial stage of drawing down, when the reforging tools completely covered the stock, until the stock diameter was less than $2\frac{1}{2}$ times the length of stock in contact with the tool and by subsequently retaining this ratio as far as possible for every stroke made near the ends of the forging.

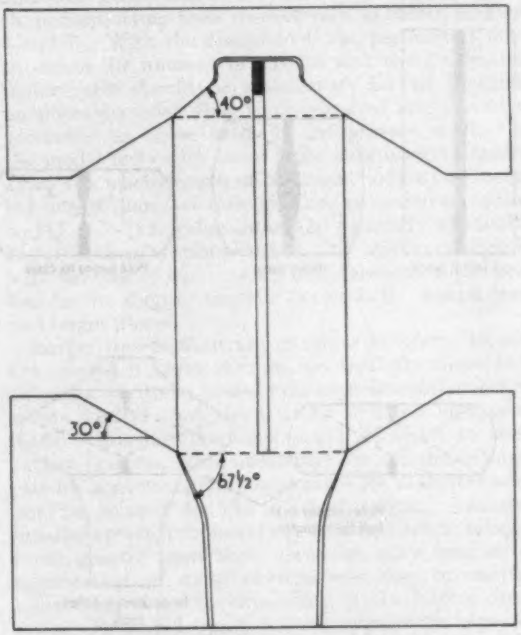


Fig. 13—Modified upsetting tools

One method of reforging was to use a pair of flat tools and a second method, recommended by Mr. R. C. Benson, employed a flat-top tool and 120° **V**-bottom tool. With this second method, a suitable angle of rotation between passes was found to be just over 50°. This ensured that previously forged faces were presented parallel to the working faces of the tools in later passes. The results of reforging using the **V**-bottom tool are illustrated in Fig. 14.

DISCUSSION

Mechanism of Flow

At present there is no theory available to explain why the central holes should close up when compress-

Table II

MECHANICAL PROPERTIES OF SOLID-PHASE WELDS

	Test No.	Yield Point, tons/in²	Max. Stress, tons/in²	Elongation, %	Reduction in Area, %	Remarks
Material welded during upsetting and reforged	1	18·4	33·0	33½	54½	The fractures did not indicate the presence of a weld
	2	18·4	33·6	31½	57	
	3	17·6	33·4	29	50	
Material welded during upsetting	4	18·8	30·8	26	34	These test pieces broke at the position of welding
	5	18·8	30·7	16½	21	
Original bar after upsetting	6	18·8	30·9	31	55	
	7	...	30·9	26½	50	

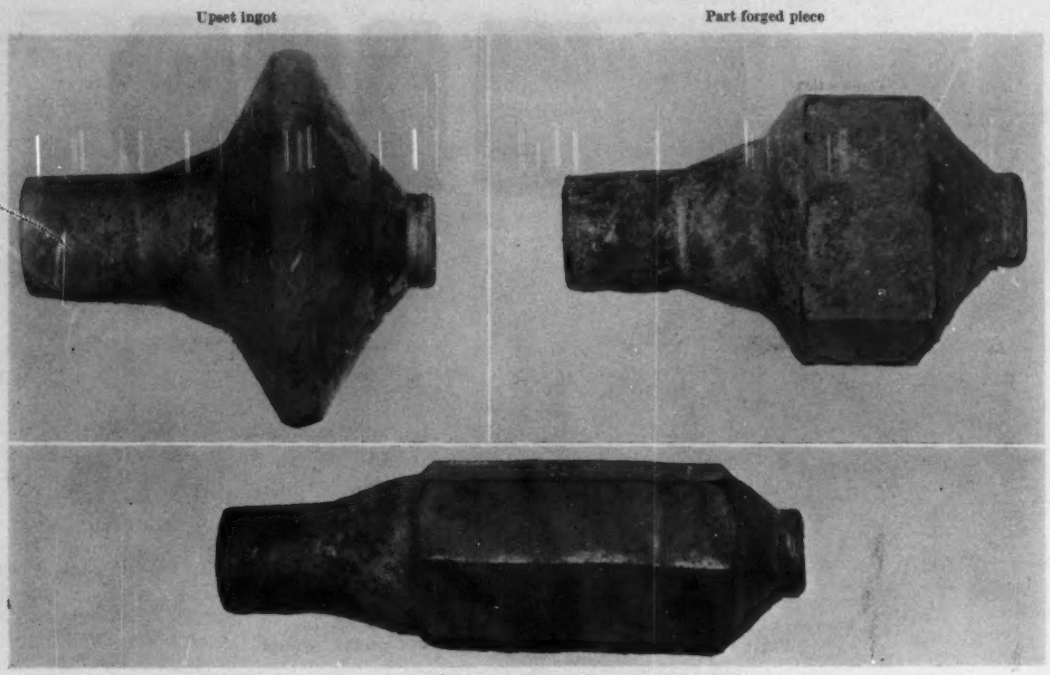


Fig. 14—Reforging tests

ing what are essentially thick-walled cylinders, or to account for the relationship between the portion of hole closed and the height/diameter ratio of the compressed piece at any instant.

While the present work has been concentrated mainly on the practical aspects of the upsetting operation, a number of points have been noted which may have a bearing on the mechanism of closure. These suggest that the results of upsetting, which are illustrated diagrammatically in Fig. 15, could be ascribed to the frictional effects which exist at the contact faces between the tools and the test block.

The friction, which is sufficient to prevent movement at the contact faces, imposes a constraint on movement varying throughout the material. Regions of high constraint are present in the vicinity of the tool faces, and in tall blocks ($H/D > 1$) these regions remain undeformed; this explains why holes near the ends of tall blocks remain unaltered in shape. Such regions have previously been postulated for other types of forging operations and the usual conception is that they are cone-shaped. On this assumption, measurement of the length of the undeformed portions of full-length holes, in blocks of circular section with H/D ratios between 1 and 2 indicates that the base angle of the cones was in the region of 30°.

Away from the ends the frictional constraint is less and the deformation of a hole in the central region of the block follows the same trend as that of the block as a whole, i.e. it is increased in diameter and reduced in height.

As upsetting is continued, the two regions of high constraint at each end of the block converge. It appears that the holes begin to shrink in diameter when the tips of the two cones meet. After further upsetting the extent of the hole shrinkage appears to coincide with the overlap of the two constrained regions.

The effect of dishing the working faces of the up-

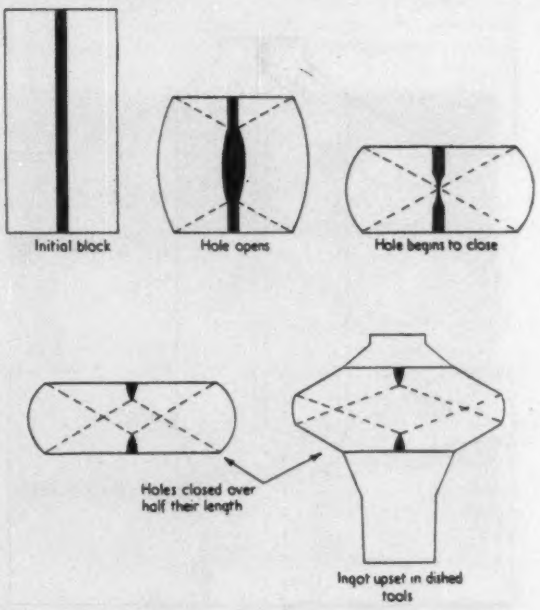


Fig. 15—Diagram of effects during upsetting

setting tools is, apparently, to enlarge the regions of frictional constraint.

Application of Results

The experiments have indicated conditions in which upset forging will close artificial defects on or near the axis of test blocks, and the nature of the results suggests that this is also likely to occur with cavities in forging ingots. The artificial holes were closed because material flowed inwards from all directions to fill them. This type of flow, which appears to be associated with the presence of friction between the stock and the forging tools, would exist in industrial upsetting operations and should also ensure the complete closure of any irregularly shaped cavity which might occur in practice.

Inward flow first occurred at mid-height of the stock and with continued upsetting it spread towards the ends. The distance over which the test holes were healed depended on the final shape of the upset piece, as defined by the height/diameter ratio, regardless of the initial shapes of the stock, which were chosen to cover the normal range of ingot shapes used in forging. This finding, which is a new concept for upsetting practice, indicates that it would be preferable to measure the height/diameter ratio during upsetting instead of using the conventional upsetting ratio which measures the reduction in height. For example, an ingot with a 30-in. dia. and 60-in. long body would require an upset in the dished tools of about 2·8 : 1 to achieve the desired final shape, whereas an ingot of the same volume with a 38-in. dia. and 38-in. long body, which would have the same final shape, would require an upset of 1·75 : 1.

The relationship between the amount of closure and the height/diameter ratio, for flat and dish-shaped upsetting tools respectively, is shown in Figs. 4 and 7. With the dished tools the parameters used to define the amount of closure and the height/diameter ratio should be satisfactory for all practical purposes provided that the tools used are of similar geometry to those used in the present work. In the model tests with these tools, cavities were healed along the whole length of the ingot body by reducing the height/diameter ratio to 5 : 8, measured according to Fig. 7. This value should be generally applicable to full-scale operations, although it may vary slightly with the size of ingot owing to differences in the heat loss to the forging tools as between the model tests and larger pieces.

Larger upsets than are common in heavy forging are needed if ingot cavities, particularly those near the ends, are to be healed and experimental measurements suggest that these would demand high press loads. This disadvantage would be offset to some extent because, with upsetting, the weight of ingot can be kept to a minimum since its diameter need not be related to the finished forging diameter. Smaller upsets, producing pieces whose height is equal to or greater than their diameter, may lead to an aggravation of axial unsoundness due to cavities away from the ends expanding in the lateral directions. Repeated upsets of this type, with repeated reforging to the original size, are likely to result in some cavities expanding and then contracting during reforging without positively closing, although other cavities would flatten and weld during the upsetting operation. These effects stress the need for producing the relatively low height/diameter ratios which ensure that the upset pieces are free from cavities.

The material which was mechanically tested was healed under conditions as severe as any likely to be encountered industrially and all the healed material had good ductility and yield strength after the upset blocks had been reforged. The lower ductility of welded material after upsetting only was most probably due to the artificial conditions of the tests, which resulted in oxide films at the weld interfaces.

The 30° dish-shaped upsetting tools provide a means of obtaining upset pieces which are free from cavities and which at the same time can be reforged without lapping over at the ends. The dimensions of the model tools were based on the volume of the smallest ingot used which would completely fill the dished portions when the tools met. However, when the height/diameter ratio is reduced to 5 : 8 this volume does not fill the tools, and ingots up to $1\frac{1}{2}$ times the minimum volume can be accommodated without overfilling. The reforging procedures indicated in the experimental work should be applicable to most industrial forging.

CONCLUSIONS

The experiments suggest that, by using an upsetting operation before final forging down, it should be possible to produce forgings in which all internal cavities have been healed, the healed regions having good mechanical properties.

The results introduce a new concept to upset forging whereby the closure of cavities bears a definite relation to the final shape of the upset piece, irrespective of the initial shape of the stock. To produce upset pieces free from cavities, larger upsets than are common in heavy forging would be necessary, requiring high press loads. Smaller amounts of upsetting, even if repeated, are likely to cause an aggravation of axial unsoundness.

The upsetting tools should be of the dished shape indicated in Fig. 13, if the pieces are to be reforged without lapping over at the ends. With these tools, upsetting should be continued until the height/diameter ratio reaches 5 : 8, as measured in Fig. 7, although in the case of forgings which require only a central portion of the axis to be free from cavities the appropriate height/diameter ratio can be obtained from Fig. 7. The reforging procedure should be designed to avoid lapping over at the ends of the forging and a suitable method is described in the experimental work.

Acknowledgments

The authors wish to express their appreciation of the invaluable help given by the members of the Forging Committee during the course of this investigation, and particularly to Mr. H. Hattersley (Steel, Peech and Tozer), who suggested the line of work, and to Mr. R. C. Benson (The Darlington Forge Ltd.), who designed the dished tools. They would also thank their colleagues, especially Mr. A. R. Hardy, for their painstaking experimental work.

The Effect of Phosphorus on the Tensile and Notch-impact Properties of High-purity Iron and Iron–Carbon Alloys

By B. E. Hopkins, M.Sc., and H. R. Tipler, B.Met.

SYNOPSIS

Part I of this paper deals with the tensile and notch-impact properties at temperatures spanning the tough to brittle transition of Fe–P alloys for three rates of cooling. The effect of carbon in counteracting the severe intergranular weakness induced by phosphorus is dealt with in Part II, together with some results on the effect of manganese in a composition simulating that of a mild steel. The characteristics of the intergranular weakness as developed by isothermal treatments are considered in Part III, in which a comparison is made between the effects obtained in an Fe–P–C alloy and those encountered with temper brittleness in steels. An explanation of the intergranular brittleness is put forward in terms of segregation of phosphorus to the grain boundaries, which seems to be justified by the results of a separately reported parallel investigation in which experimental evidence for this segregation was obtained. Carbon is also known to be capable of segregating at grain boundaries, and so could largely prevent the phosphorus doing so. Attention is drawn to the approximately linear rise with testing temperature of the brittle fracture stress of severely embrittled Fe–P alloys, a behaviour that is in contrast to the temperature-dependence of cleavage strength.

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THE WORK REPORTED HERE on the effects of phosphorus on high-purity iron was undertaken as part of a programme that has been followed for some years on the influence of alloying elements on the mechanical properties of ferrite, with particular reference to the transition from a tough to a brittle fracture. In common with elements such as oxygen and nitrogen, the effects of which have been investigated previo usly,[1, 2] phosphorus was found to have a pronounced effect in weakening the grain boundaries of iron. In view of this and of the occurrence of phosphorus in commercial steels, it was decided to extend the work to Fe–C alloys to determine whether the deleterious effects of phosphorus in iron were modified by the presence of carbon. A still closer approach to a mild-steel composition was obtained by adding manganese.

It had been found earlier[2] that very severe grain-boundary weakness could be developed in Fe–N alloys by isothermal treatments in the approximate temperature range 500–800° C. The development of intergranular brittleness in the alloys dealt with in this paper was consequently also investigated by isothermal heat-treatments; the modification of its characteristics by the presence of carbon left little doubt of a direct connection with temper brittleness in commercial steels.

The results are reported in three sections: the effect of phosphorus in iron, the modification of this by the presence of carbon and carbon plus manganese, and the development of intergranular brittleness during isothermal heat-treatment.

Manuscript received 5th July, 1957.
The authors are at the National Physical Laboratory.

PREPARATION OF ALLOYS

Two series of alloys were made using commercial electrolytic iron of Japanese or Norwegian origin as the starting material. The alloys were made by the method described by Hopkins, Jenkins, and Stone[3], which consists of melting first in air to oxidize impurities such as carbon, silicon, and manganese, and then in hydrogen at atmospheric pressure to reduce the oxygen and nitrogen contents to low levels. As the base irons had an appreciable sulphur content (0·01–0·03%), during the melting in air a desulphurizing slag treatment was applied to the iron under reducing conditions before the oxidizing treatment, as described below. The melting in hydrogen was done in either a 25-lb or a 50-lb high-frequency furnace.

Phosphorus additions were made in two ways. In the first series of binary Fe–P alloys, a master alloy containing 0·6% P, made previously by adding red phosphorus to molten iron that had been deoxidized with hydrogen, was used as the addition agent after deoxidation of the iron in the hydrogen furnace. For the second series, consisting of binary Fe–P and ternary Fe–P–C alloys, phosphorus was added as iron phosphate containing about 20% P and made by dissolving high-purity iron in phosphoric acid. When the iron phosphate was added to iron in the hydrogen furnace, complete reduction occurred with substantially no loss of phosphorus. For the larger phosphorus contents, the bulk of phosphate required was inconveniently large for addition in the hydrogen furnace. It was found that phosphorus was absorbed into the iron if the phosphate was added to the oxidized melt just before casting but the yield was only about 30%. Nevertheless, the method was otherwise

satisfactory and was used for the higher phosphorus levels in the second series of alloys. The mechanism of absorption of phosphorus in the oxidized melts is presumably straight solution of the phosphate accompanied by oxidation of the iron.

The carbon for the Fe–P–C alloys was added as Hilger H.S. electrodes to the hydrogen-reduced iron. It was necessary to pour the melt immediately the carbon was absorbed so as to minimize contamination by aluminium resulting from reduction of the alumina melting crucible.

The effect of adding 0·5% Mn to one Fe–P–C alloy was investigated. The manganese used was electrolytic in origin; it was added to the melt after deoxidation of the iron by hydrogen, but, as it contained some oxygen, this was removed by prolonging the hydrogen treatment of the melt.

The desulphurization treatment applied at the air-melting stage as mentioned above consists of deoxidizing the melt with aluminium, adding 0·5% C to maintain reducing conditions, and then treating with a 1:1 lime–alumina mixture with sufficient soda ash to give a fluid slag. All the materials used have to be low in sulphur. The slag is left on for about $\frac{1}{4}$ h and skimmed off. Because the volume of slag which can be used at a time is small, it is necessary to repeat the treatment about six times to reduce the sulphur content below 0·005%. The melt is then allowed to oxidize to remove the aluminium and carbon that were added in this treatment as well as other impurities present initially in the iron.

The alloys made as above were chill cast under a small pressure of hydrogen to give 25-lb or 50-lb ingots which were hot-rolled to $\frac{5}{8}$-in. dia. bar. Analyses of the alloys are given in Tables I and III.

TESTING AND HEAT TREATMENT

Tensile tests at room temperature, — 78° C, and — 196° C were carried out on specimens 0·282 in. dia. with a 1-in. gauge length. Proof stresses were determined by means of an extensometer[4] that is not damaged if a test-piece breaks before these stresses are reached. Charpy impact tests on specimens 10 mm square by 55 mm long with an Izod V-notch (2 mm deep, 0·25 mm root radius) were done at various temperatures covering the tough to brittle transition. The transition temperature in the impact tests has been taken as the temperature at which the impact value was half that in the fully tough condition.

Tensile and impact tests were carried out initially on the alloys in three conditions of heat-treatment, viz. furnace-cooled, air-cooled, and water-quenched from 950° C. The furnace-cooling rate was very slow, the average rate from 950–700° C being 1° C/min with a total time of 4½ days to reach room temperature. Further investigation of the intergranular brittleness to which most of the alloys were found to be susceptible involved tensile tests, mainly at — 196° C, and some Charpy impact tests on specimens heat-treated isothermally for certain periods of time at temperatures in the range 300–1100° C followed by quenching in water. These treatments were usually carried out in evacuated silica or Pyrex tubes, but some short-time treatments were done by quenching from a lead or salt bath.

Table I

ANALYSES OF JAPANESE ELECTROLYTIC IRON-BASE Fe–P ALLOYS

Element, %	84AF1 Fe	85AF1 Fe–0·044%P	87AF2 Fe–0·15%P	82AF2 Fe–0·31%P
Carbon	0·003	0·003$_5$	0·001$_5$	0·002$_0$
Silicon	0·005	0·001	0·003	0·001
Manganese	0·002	<0·001	n.d.	n.d.
Sulphur	0·006	0·004	0·003	0·004
Phosphorus	0·001	0·044	0·15	0·31
Nickel	n.d.	0·005	0·005	0·006
Chromium	n.d.	0·001	n.d.	0·002
Copper	n.d.	0·003	0·006	0·004
Aluminium	n.d.	0·004	0·002	0·002
Oxygen	0·002	0·0008	0·001	0·0007
Nitrogen	0·0009	0·0008	0·002	0·002
Hydrogen	Trace	Trace	Trace	Trace

n.d. = not determined
The arsenic content of alloys made from this base iron is normally 0·001%

The Charpy impact test-pieces were notched after heat-treatment. Possible quench-ageing effects in the air-cooled and water-quenched material were avoided by testing soon after heat treatment.

PART I—IRON–PHOSPHORUS ALLOYS

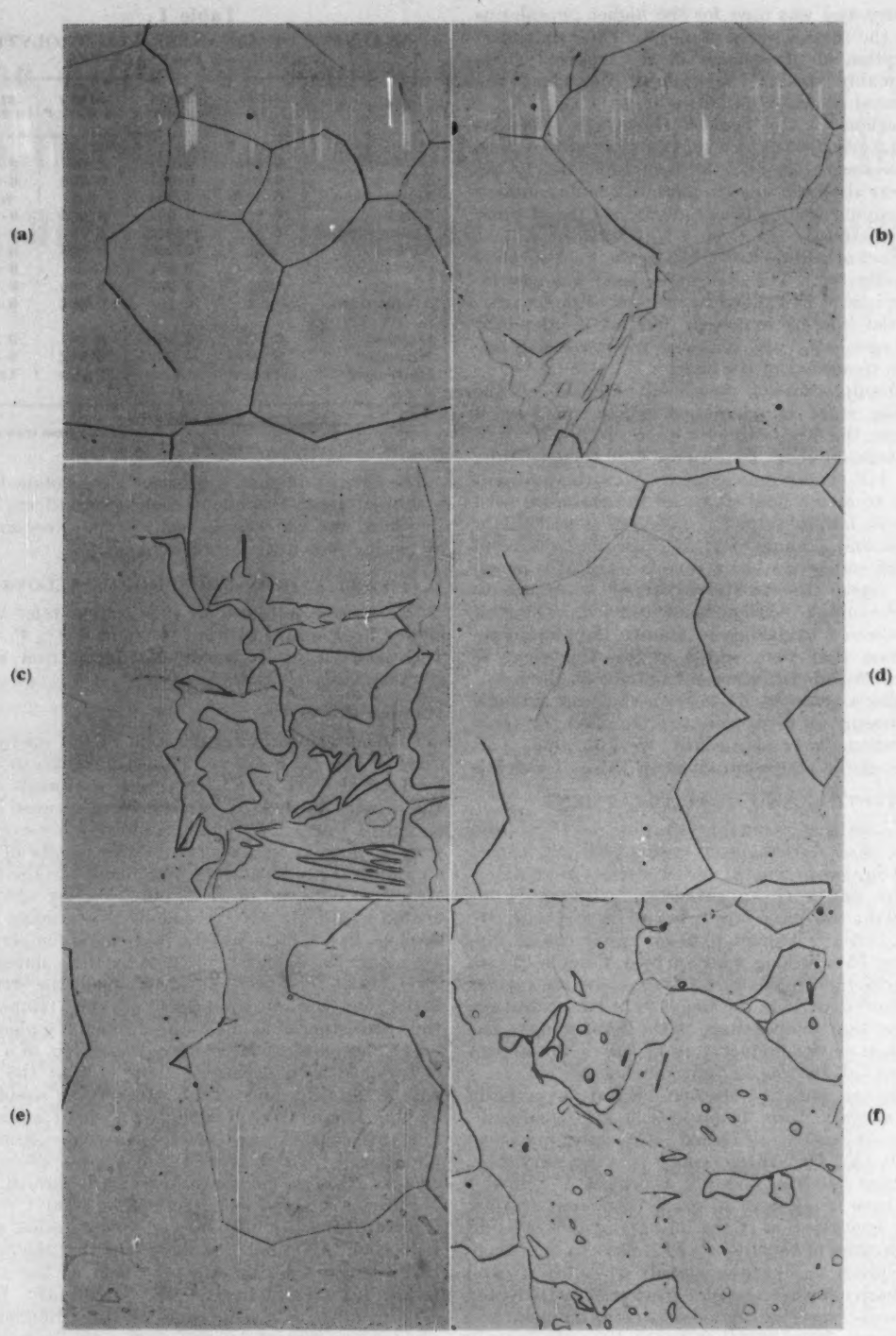
The results described in this section refer to the binary Fe–P alloys with up to about 0·3% P, which were made using Japanese electrolytic iron as the starting material (Table I).

Microstructures

To facilitate the interpretation of the microstructures of the Fe–P alloys in the conditions in which mechanical tests were carried out, specimens of the alloys were heated to various temperatures in the region of the γ loop and examined microscopically after cooling at various rates. The results of these experiments are illustrated by reference to the microstructures obtained in the 0·146% P alloy which was heated at 935° or 1000° C followed by cooling in the furnace, in air, or in water. Both these temperatures are within the duplex $\alpha + \gamma$ field for this composition: 935° C is at the lower end of this field, the structure under equilibrium conditions at this temperature thus consisting of a small amount of the γ phase of a lower phosphorus content than the matrix of α which is enriched in phosphorus; 1000° C is at the upper end of the field, and so the structure at equilibrium at this temperature consists of a small amount of untransformed α phase which has become enriched in phosphorus, in a matrix of γ of a lower phosphorus content than the ferrite from which it formed.

Furnace cooling from either 935° or 1000° C resulted in microstructures of polygonal ferrite grains, and no indication was found in the structure that the material had been in the duplex $\alpha + \gamma$ field at the soaking temperature (Fig. 1*a* and *d*). Apparently, furnace cooling gave sufficient time for the differences in phosphorus contents of the α and γ phases present at the soaking temperature to be evened out by diffusion.

The structures obtained both by air cooling and water quenching were duplex in appearance as a result of the presence of the α and γ phases at 935° and



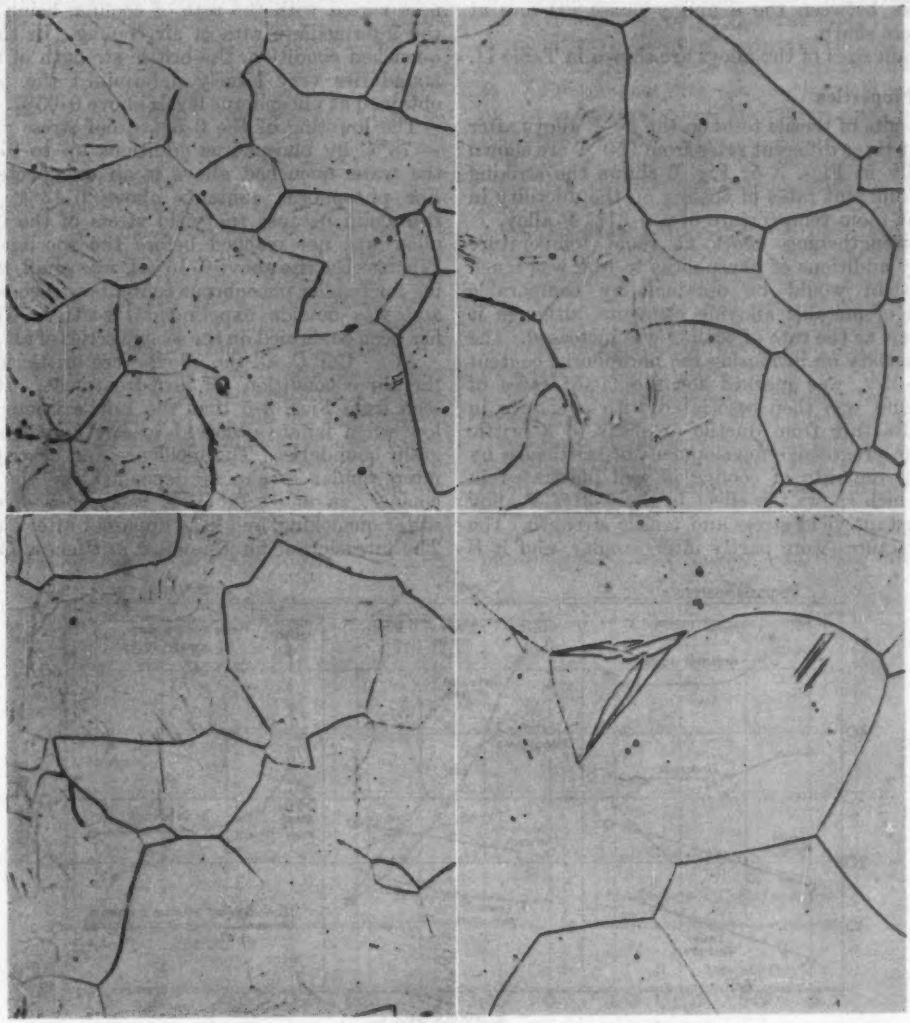
(*a*) Furnace-cooled 935° C
(*b*) Air-cooled 935° C
(*c*) Water-quenched 935° C
(*d*) Furnace-cooled 1000° C
(*e*) Air-cooled 1000° C
(*f*) Water-quenched 1000° C

Fig. 1—Microstructures of Fe–0·146%P alloy after various heat-treatments × 150

1000° C. In the air-cooled specimens (Fig. 1*b* and *e*), the boundaries between the areas of α and γ at the soaking temperature were indistinct, suggesting that etching revealed the two types of area because of differences in phosphorus content which persisted due to insufficient time for homogenization during cooling, rather than because of differences in orientation. That there were differences in phosphorus content was shown by etching in acid copper chloride solution. In the water-quenched specimens (Fig. 1*c* and *f*), the boundaries between the areas of α and γ at the soaking temperature were sharp, suggesting differences in orientation as well as differences in phosphorus content. On air cooling, the γ regions appeared to have transformed to α by attaching themselves to existing α, whereas on water quenching the γ regions appeared to have adopted their own orientation when transforming to α.

In the structures of the specimens heated at 935° C, it can be seen that the γ phase formed mainly at the grain boundaries of the α phase, so that, apart from effects of any grain growth that might have occurred, the initial grain boundaries remained after cooling. At 1000° C the γ phase grew into the α grains and represented the major portion of the specimen, so that the grain boundaries would not necessarily be the same as before treatment.

The above enables the microstructures of the alloys of various phosphorus contents, in the conditions of heat-treatment for which mechanical properties were determined, to be readily understood. These treatments were furnace cooling, air cooling, and water quenching from 950° C. In the furnace-cooled condition, all the alloys, which had phosphorus contents of 0·05–0·31%, had microstructures consisting of polygonal ferrite grains. Figure 2*a–d*, showing the structures after air cooling from 950° C, indicate that at this temperature all the alloys were



(*a*) Fe-0·05%P (*b*) Fe-0·086%P (*c*) Fe-0·146%P, (*d*) Fe-0·31% P

Fig. 2—Microstructures of Fe–P alloys air-cooled from 950° C × 150

Table II—GRAIN SIZES OF ALLOYS LISTED IN TABLE I

Alloy	Grain Size					
	F.C. 950° C		A.C. 950° C		W.Q. 950° C	
	A.S.T.M.	grains/mm	A.S.T.M.	grains/mm	A.S.T.M.	grains/mm
84AF1 Fe	2–3	6·9	2–3	6·3	2	5·8
85AF1 Fe–0·044%P	3	8	4	10·3	0	2·8
87AF2 Fe–0·15%P	0	2·6	2–3	7	<0	2·1
82AF2 Fe–0·31%P	0	2·5	1–2	4·7	1	3·5

within the duplex $\alpha + \gamma$ field but that the proportion of α increased progressively as the phosphorus content was raised, the 0·05% P alloy being near the $\gamma/\alpha + \gamma$ boundary and the 0·31% P alloy near the $\alpha + \gamma/\alpha$ boundary. After water quenching, the structures were similar to those after air cooling except that the boundaries between the α and γ phase existing at 950° C were sharp.

The grain sizes of the alloys are shown in Table II.

Tensile Properties

The results of tensile tests on the Fe–P alloys after cooling at three different rates from 950° C are shown graphically in Figs. 3–5; Fig. 6 shows the striking effect of different rates of cooling on the ductility in tension at room temperature of a 0·31% P alloy.

The strengthening effect at room temperature caused by additions of phosphorus to iron was much greater than would be obtained by comparable additions of metallic alloying elements, although it became less as the rate of cooling was increased. The fall in ductility on increasing the phosphorus content above 0·05% was marked for the faster rates of cooling, and was then associated with a change in type of fracture from ductile or shear to a brittle type. The progressive development of brittleness by increasing the rate of cooling is well illustrated in Fig. 6, which shows the effect in the 0·31% P alloy for a constant yield stress and tensile strength. The brittle fractures were partly intergranular, and it is apparent that this feature became increasingly important in determining the properties as the temperature of test was lowered.

At — 78° C, the hardening effect of phosphorus was partly masked by the rapid onset of brittleness in alloys with more than 0·05% P; the degree of brittleness varied with the rate of cooling, being least for the intermediate rate of air cooling. In the water-quenched condition, the brittle strength of the grain boundaries very largely determined the properties obtained at phosphorus levels above 0·05%.

The lowering of the 0·5% proof stress of iron at — 78° C by phosphorus additions up to 0·15% for the water-quenched alloys is an unexpected effect. For phosphorus contents above 0·15%, it must rise again because the yield stress of the 0·31% P alloy was not reached before the specimen broke. Whereas the rise above 0·15% P was small, the lowering for smaller phosphorus contents was considerable, and was outside experimental scatter. The effect has been confirmed on the second series of alloys.

At — 196° C, all the alloys were brittle in each of the three conditions of heat-treatment. It will be seen from Figs. 3–5 that the brittle strengths were low when failure occurred to any extent along the grain boundaries. The incidence and severity of the intergranular weakness depended on the rate of cooling: as at — 78° C, it was most severe after water quenching, and least apparent after air cooling. The intrusion of intergranular brittleness prevents a

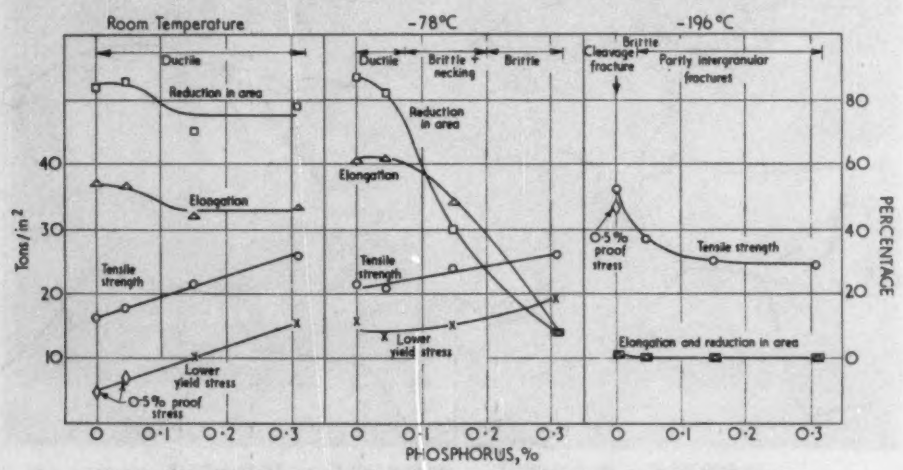


Fig. 3—Tensile properties of Fe–P alloys after furnace-cooling from 950° C

Table III

ANALYSES OF NORWEGIAN ELECTROLYTIC IRON-BASE ALLOYS OF Fe–P, Fe–P–C, AND Fe–P–C–Mn

Element, %	Mark												
	ACV	ACH	ACT	ACU	ACM	ACN	ACP	ACQ	ACK	ACJ	ACR	ACS	ACL
Carbon	0·0034	0·002	0·003	0·001	0·051	0·050	0·052	0·052	0·15	0·16	0·15	0·14	0·15
Silicon	0·006	0·002	0·005	0·002	0·004	0·002	0·001	0·005	0·007	0·004	0·004	0·004	0·004
Manganese	<0·005	0·003	<0·005	<0·005	<0·002	<0·002	<0·002	0·002	<0·002	<0·002	<0·002	<0·002	0·51
Sulphur	0·004	0·006	0·004	0·003	0·0026	0·003	0·0023	0·0046	0·003	0·007	0·002	0·002	0·003
Phosphorus	<0·001	0·049	0·086	0·146	<0·001	0·058	0·095	0·16	0·001	0·05	0·12	0·144	0·053
Oxygen	0·0006	0·0008	0·0006	0·002	0·0005	0·0005	0·0007	0·0006	0·0006	0·0008	0·0005	0·0006	0·0005
Nitrogen	0·001	0·0014	0·001	0·001	0·0005	0·0005	0·0006	0·0006	0·0006	0·0016	0·0006	0·0008	0·0007
Hydrogen	← trace →												

Nickel, chromium, copper, and arsenic were not determined in these alloys but were <0·001% in the base iron. The zinc content of alloys made from this base iron is <0·001%.

general assessment of the effect of phosphorus on cleavage strength, but for the air-cooled material, since failure occurred entirely by cleavage in the iron and 0·05% P alloy, it can be concluded that this amount of phosphorus raised the cleavage strength. In the main, however, the tensile strengths at — 196° C represented essentially the brittle strengths of the grain boundaries.

Charpy Impact Properties

The variations in transition temperature in the V-notch Charpy impact test with increase in phosphorus content are shown in Fig. 7, which also indicates the type of fracture in the specimens that broke in a brittle manner for each phosphorus level. The stage at which intergranular failure became apparent in the brittle specimens for each condition of heat-treatment was the same as for the tensile tests at — 196° C. In interpreting variations in the Charpy impact test results in terms of changes in the brittle strength and yield stress, ideally these changes should refer to temperatures in the transition range, but as this is not possible it will be sufficient for discussion of large effects on transition temperature to consider changes in the brittle strength at — 196° C and the yield stress at room temperature.

In the furnace-cooled condition, the fractures of the brittle specimens became partly intergranular with the first addition of phosphorus, and the transition temperature was raised continuously. This is consistent with the drop in brittle strength as measured at — 196° C, and with the rise in yield stress found at room temperature for increasing phosphorus contents.

In the air-cooled condition there was a small drop in transition temperature in the range of phosphorus contents that gave wholly cleavage fractures in the brittle specimens, that is up to 0·05%, but, for higher phosphorus contents, partly intergranular fractures were obtained, and the transition temperature was raised. These results are again consistent with the brittle strengths measured at — 196° C and with the effect of phosphorus in raising the yield stress at room temperature.

In the water-quenched condition, all the brittle specimens broke partly along the grain boundaries. The transition temperature went through a minimum at 0·05% P, as would be expected if the brittle strength were the dominating factor in governing the transition temperature.

Whereas the shapes of the transition-temperature/phosphorus-content curves are thus interpretable

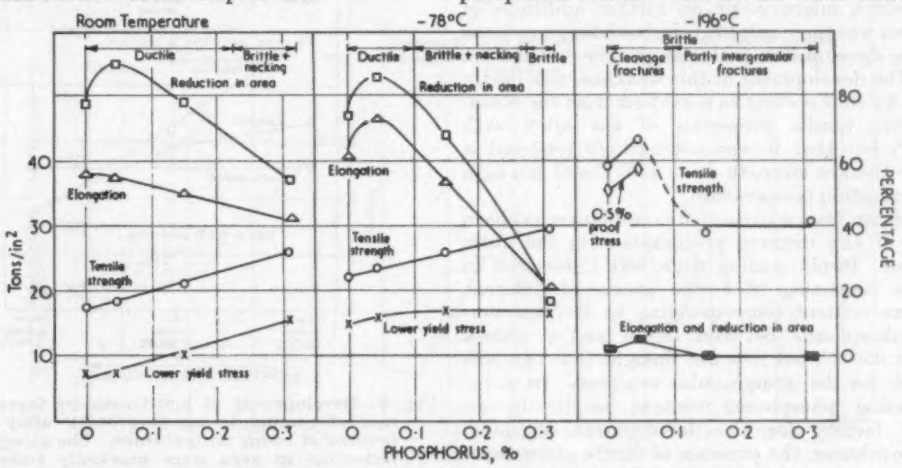


Fig. 4—Tensile properties of Fe–P alloys after air-cooling from 950° C

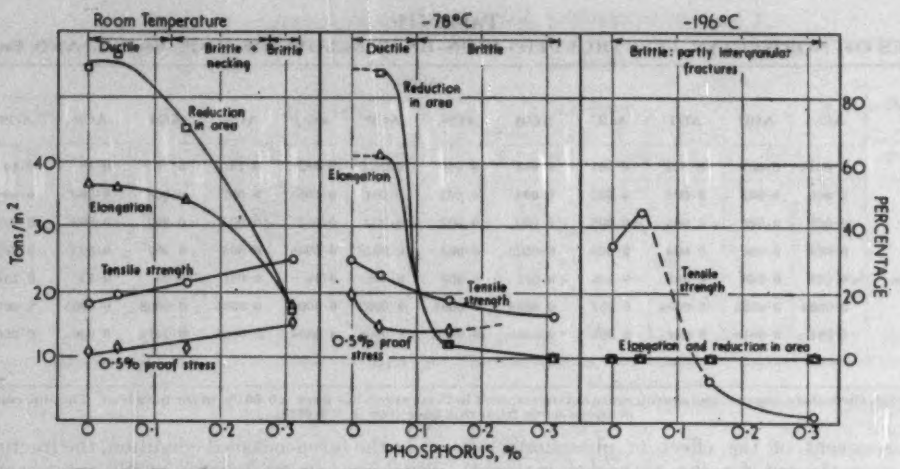


Fig. 5—Tensile properties of Fe–P alloys after water-quenching from 950° C

in terms of the brittle strength at — 196° C and the yield stress at room temperature, the relative values of the transition temperature for the different conditions of heat-treatment are only in general agreement with such considerations, there being some items of detail that are not explained in these terms. For example, the transition temperature of the 0·31% P alloy was a little lower when furnace cooled from 950° C than when air cooled, although the yield stresses at room temperature were similar and the brittle strength at — 196° C was higher for the air-cooled condition.

Consideration of Results

One of the most important effects of the addition of phosphorus to iron of high purity was an embrittlement of the grain boundaries, and the results of the tensile and impact tests of Fe–P alloys must be considered in the light of this fact. At room temperature the tensile properties were slightly improved by the addition of not more than 0·05%P but the ductility in the notched-bar impact test might be adversely affected. The rapid fall in ductility in tension which might occur on further additions of phosphorus was the result of grain-boundary weakness which was developed in material quickly cooled from 950° C. The development of this weakness was partly inhibited by slow cooling as is evident from the room-temperature tensile properties of the alloy with 0·31% P, but that it was not entirely removed is shown by the low strength at — 196° C and the high impact transition temperature.

No evidence from microscopic examination has been obtained of any discrete precipitation in the grain boundaries. Rapid cooling from 950° C resulted in structures consisting of ferrite grains of different phosphorus content (corresponding to the distribution of phosphorus between the α and γ phases present at 950° C) but it is not thought that this was responsible for the intergranular weakness. In so far as increasing phosphorus content results in an enhanced facility for the development of intergranular weakness, the presence of ferrite grains with a phosphorus content higher than the average might result in increased brittleness of the rapidly cooled material.

PART II—IRON–PHOSPHORUS–CARBON ALLOYS

In the Fe–P–C series of alloys the largest phosphorus addition was less at 0·15% than in the binary Fe–P alloys dealt with in Part I. The alloys investigated had carbon contents of 0·05% and 0·15%. The effect of manganese was determined by adding 0·5% to an Fe–0·05%P–0·15%C alloy so that the composition represented that of a high-purity mild steel.

Norwegian electrolytic iron was used in making the Fe–P–C alloys. To check that the different starting materials did not have individual effects that would make a comparison of the properties difficult, some Fe–P alloys were included in the series made from the Norwegian base iron. The properties at various temperatures of the Fe–P alloys of the second series were determined in the air-cooled condition only. From Fig. 8 it will be seen that the tensile properties were very similar for the two sets of alloys.

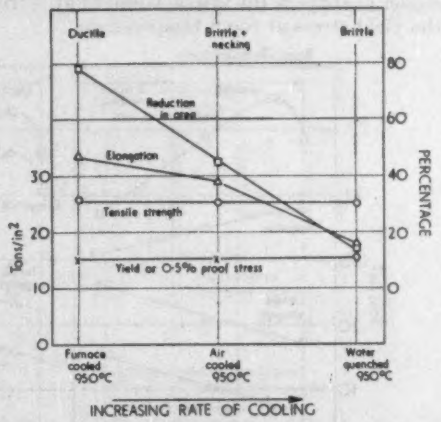


Fig. 6—Development of brittleness by increasing the rate of cooling in an Fe–0·31%P alloy tested in tension at room temperature. The elongation and reduction in area were markedly reduced for a constant yield and tensile strength

Table IV—GRAIN SIZES OF ALLOYS LISTED IN TABLE III

Alloy	Grain Size			
	F.C. 950° C		A.C. 950° C	
	A.S.T.M.	Grains/mm	A.S.T.M.	Grains/mm
ACV Fe			1–2	4·8
ACH Fe–0·05%P			4–5	15
ACT Fe–0·086%P			2–3	7·7
ACU Fe–0·15%P			3–4	9·2
ACM Fe–0·05%C	3	8·5	5–6	20·9
ACN Fe–0·05%C–0·05%P	3	9·5	6–7	25·5
ACP Fe–0·05%C–0·1%P	2	5·7	6	24
ACQ Fe–0·05%C–0·15%P	2	5·8	7	27·1
ACK Fe–0·15%C	4	10·3	7	29·4
ACJ Fe–0·15%C–0·05%P	4	12·5	7	33·6
ACR Fe–0·15%C–0·1%P	4	11·6	7	39
ACS Fe–0·15%C–0·15%P	5	18·5	7	35·9
ACL Fe–0·15%C–0·05%P–0·5%Mn	4–5	13·9	7	31·6

Microstructures

The microstructures of the Fe–0·1% and 0·15%P–0·05%C alloys and of the Fe–0·15%P–0·15%C alloy after air cooling or water quenching from 950° C indicated that these alloys had been in the duplex $\alpha + \gamma$ field associated with the γ loop during heat-treatment. In this respect these alloys had features similar to those discussed for the binary Fe–P alloys, i.e. the α phase present at the heat-treatment temperature was not detectable after furnace cooling, but had etching characteristics that revealed its presence after air cooling, and was most clearly visible after water quenching. The alloys of lower phosphorus content than those given above for each carbon level were in the γ field at 950° C. These results are consistent with a shift of the γ loop to higher phosphorus levels with increasing carbon content as indicated by Rocquet and Jegaden.[5]

In the furnace-cooled and air-cooled conditions the carbide appeared as pearlite and as films either at ferrite grain boundaries or around areas of pearlite. The grain size was somewhat finer in the higher-carbon alloys, but changing the phosphorus content had little effect on the grain size. The carbide and the grain size were both finer in the air-cooled than in the furnace-cooled condition. The grain sizes of the alloys are given in Table IV. The microstructure of the Fe–0·15%P–0·05%C alloy in the air-cooled condition is reproduced in Fig. 9*a* and *b*, which show the duplex ferrite and the form of the carbide.

The matrix structures of the alloys in the water-quenched condition varied in such a way that it was obvious that they were dependent on the precise conditions of cooling. In some cases the structure varied from end to end of a specimen, and from specimen to specimen within a batch quenched at

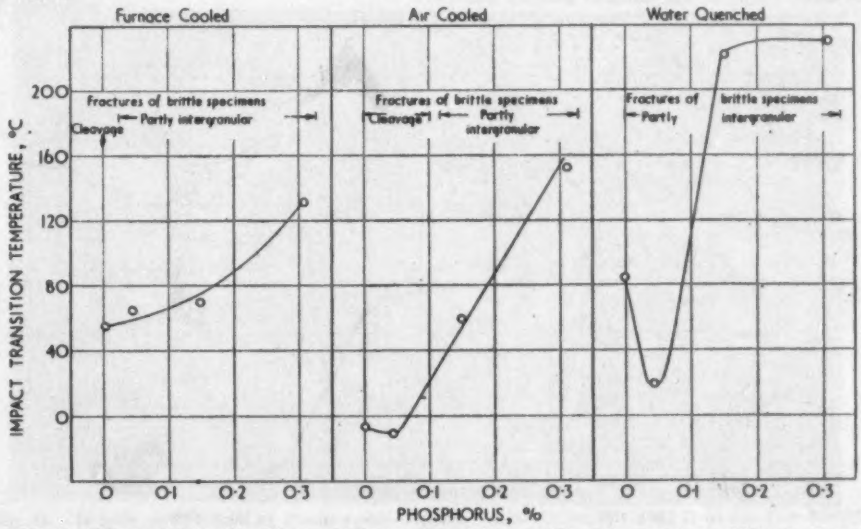


Fig. 7—Effect of rate of cooling on Charpy impact transition temperature of Fe–P alloys

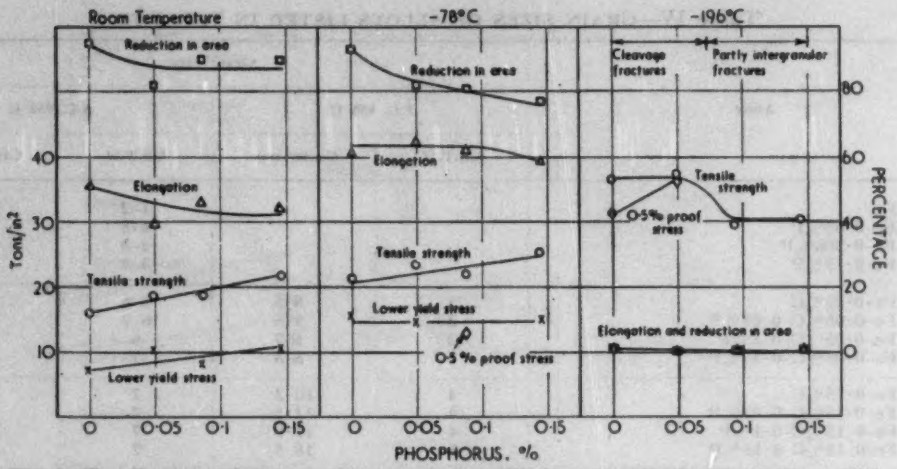


Fig. 8—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P alloys *ACV*, *ACH*, *ACT*, and *ACU* air-cooled from 950° C

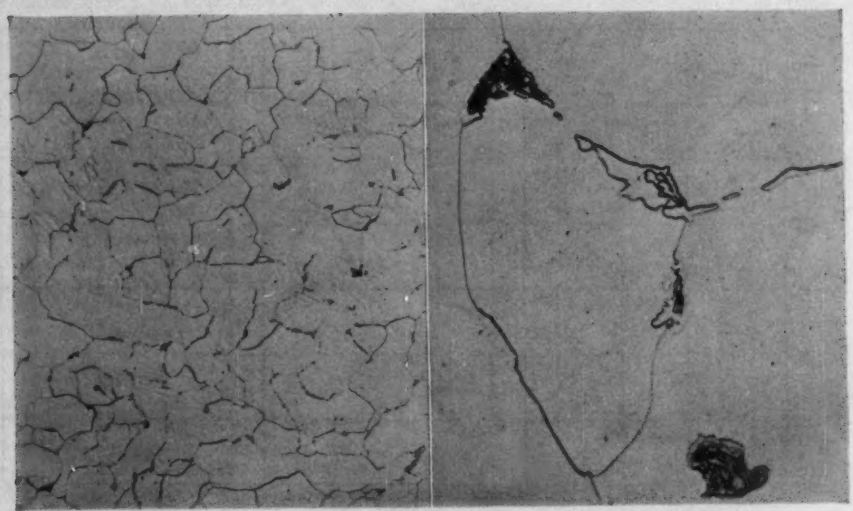
the same time. The structures usually consisted of acicular low-carbon martensite, or of ferrite formed at a somewhat slower rate of cooling and in which there was only a little evidence of carbide, or of mixtures of these. With a particularly slack quench, fine-grained ferrite and unresolvable pearlite were obtained. The more or less clear-etching ferritic structure formed at an intermediate rate of cooling is the one that has been shown previously[4] to give a characteristically low impact transition temperature.

The variation in microstructure on water quenching was more pronounced in the higher-phosphorus alloys; this could probably be accounted for by segregation of carbon as well as phosphorus since these particular alloys showed some banding. In view of the marginal hardenability, the properties in the water-quenched condition were investigated in full at the 0·05% carbon level only. The adoption of special precautions such as quenching individually from a vertical furnace would be expected to have given better uniformity of structure.

Tensile Properties

The tensile properties of the Fe–P–0·05%C alloys at various temperatures in the three conditions of heat-treatment are shown in Figs. 10–12, and of the Fe–P–0·15%C alloys for the furnace-cooled and air-cooled conditions in Figs. 13 and 14.

At 0·05% C, the hardening effect of phosphorus apparent at room temperature became much less marked or even disappeared on lowering the temperature to — 78° C. In the water-quenched condition, the scatter in properties at — 78° C was caused by the variations in microstructure mentioned above. The rapid drop in ductility at — 78° C for the furnace-cooled condition was accompanied by a change in



(*a*) × 150 (*b*) × 1500

Fig. 9—Fe–0·15%P–0·05%C alloy air-cooled from 950° C

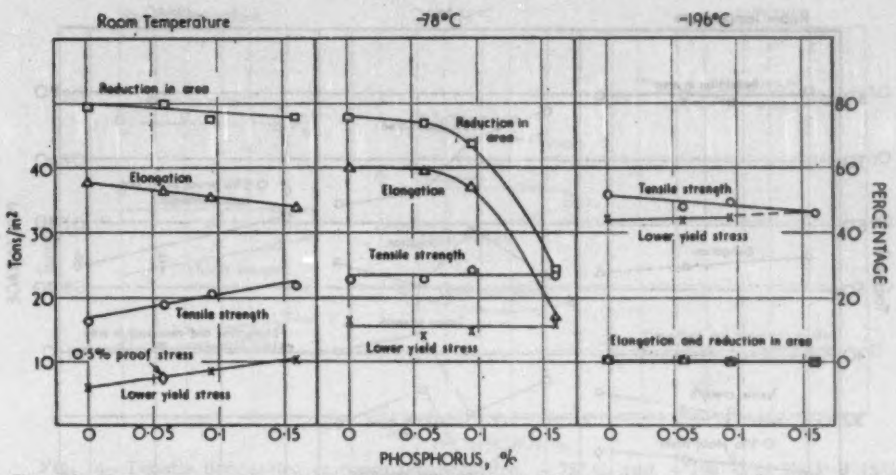


Fig. 10—Tensile properties at room temperature, — 78°C, and — 196° C of Fe–P–0·05%C alloys furnace-cooled from 950° C

type of fracture from ductile to brittle, showing that the transition temperature range in tension was raised to some extent by increasing phosphorus content. This was also apparent at — 196° C for this condition of heat treatment as shown by the alloy with the highest phosphorus content failing in a brittle manner before reaching the yield stress. At — 196° C, the alloys were almost completely brittle for the three rates of cooling. There was little effect of phosphorus on the properties at this temperature for the furnace-cooled and air-cooled conditions, except for a slight tendency for the brittle strength to be lowered as the phosphorus content was raised; the fractures, however, appeared to be entirely of the cleavage type. The lowering of the brittle strength at — 196° C with increasing phosphorus content was a little more marked for the water-quenched condition but there was no very definite evidence of intergranular failure.

At 0·15% C, the tensile properties of the alloys in the furnace-cooled and air-cooled conditions were similar to those for the 0·05% C level, but there were some small differences. The most marked of these was that in the furnace-cooled condition the fractures were ductile at — 78° C for all phosphorus levels so that the ductility remained high at the highest phosphorus content. This difference could have been due to 0·05% C being insufficient to counteract completely the embrittling effect of 0·15% P. There was also a consistent trend in the yield stress values at — 78° C for both the furnace-cooled and air-cooled alloys; the small lowering of the yield stress at intermediate phosphorus contents followed by a rise at the highest phosphorus content is similar to the effect noted earlier in the binary Fe–P alloys in the water-quenched condition. Such effects must arise from small differences in the temperature-dependence of the yield stress, i.e. the yield-stress/temperature curves for the various alloys must cross as the temperature is lowered.

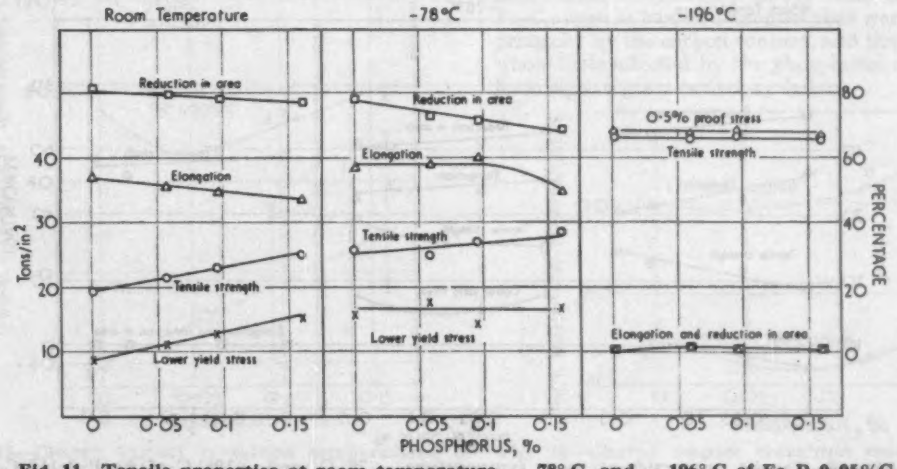


Fig. 11—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P–0·05%C alloys air-cooled from 950° C

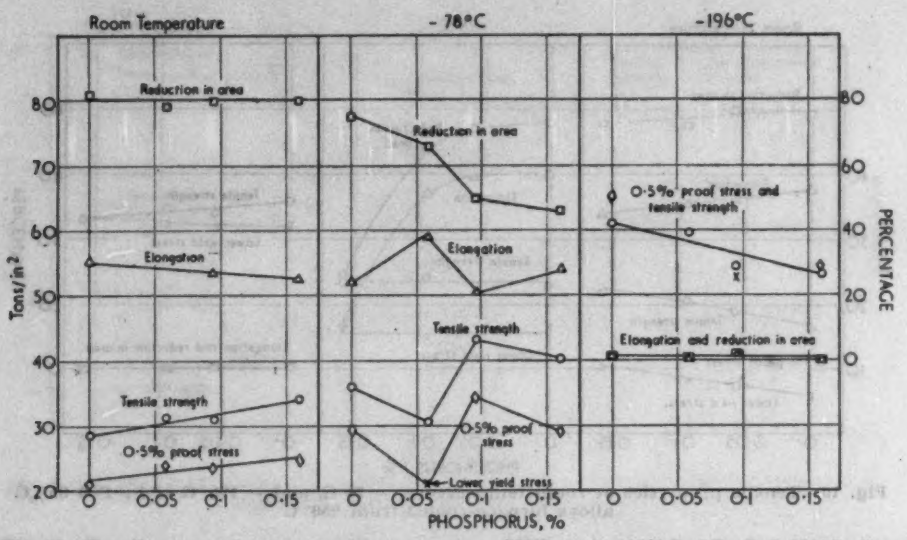


Fig. 12—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P–0·05%C alloys water-quenched from 950° C

Charpy Impact Properties

The temperatures of the transition from tough to brittle fracture in the V-notch Charpy impact test of the ternary Fe–P–C alloys after cooling at various rates from 950° C are shown in Figs. 15 and 16. The effect of phosphorus was comparatively small for the furnace-cooled and air-cooled conditions, but there was a general tendency for a minimum to appear in the curves at around 0·05% P. As this trend was not very pronounced, it is not surprising that it could not have been predicted from the yield-stress values at room temperature and the brittle strengths at — 196° C. The scatter in the results for the water-quenched condition at the higher phosphorus levels in the 0·05% C series has already been referred to in connection with the microstructures and tensile properties; nevertheless, the results indicated that for this condition there was a more consistent effect of phosphorus in raising the transition temperature. It will be recalled that there was a somewhat more pronounced drop in the brittle fracture stress at — 196° C of these alloys in the water-quenched condition.

The transition temperatures were lowered by increasing the rate of cooling. The difference between the transition temperatures of the furnace-cooled and water-quenched alloys in the 0·05% C series was not less than 120° C when the phosphorus contents were within commercial limits.

Comparison with Fig. 7 shows that in the furnace-cooled and air-cooled conditions the transition temperatures of the ternary Fe–P–C alloys were higher than those obtained in the binary Fe–P alloys. The fractures of the brittle specimens were similar to those

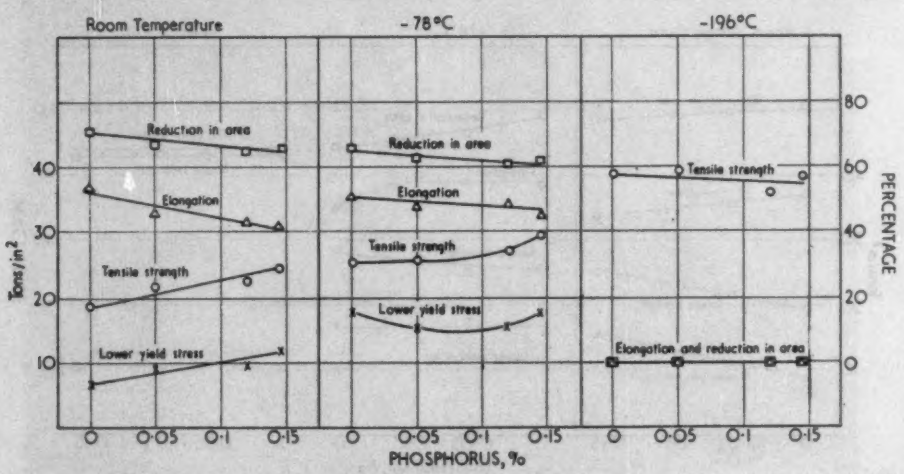


Fig. 13—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P–0·15%C alloys furnace-cooled from 950° C

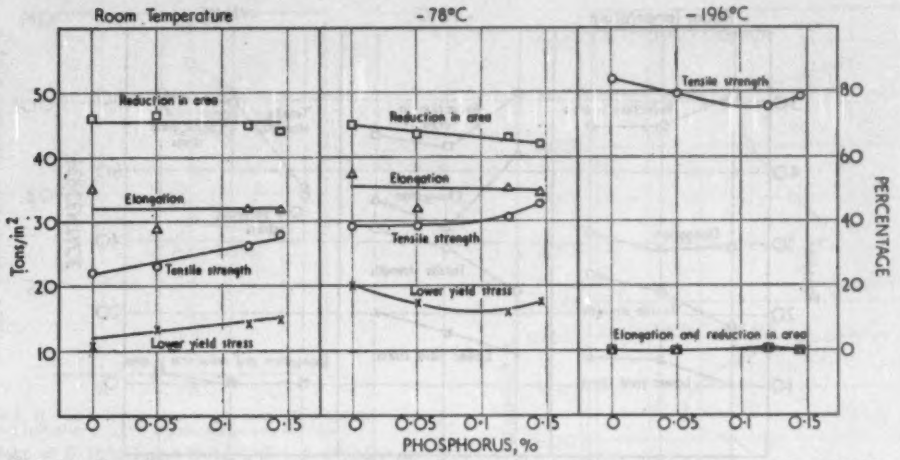


Fig. 14—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P–0·15%C alloys air-cooled from 950° C

of the tensile specimens tested at — 196° C in showing no very definite evidence of intergranular failure.

The Effect of Manganese

The alloy made to simulate a mild-steel composition by adding 0·5% Mn at a phosphorus content of 0·05% and a carbon content of 0·15% was tested in the air-cooled and furnace-cooled conditions. Its tensile properties and transition temperature in impact in the air-cooled condition are compared with those of two other alloys in Figs. 17 and 18. The manganese caused a small increase in yield stress and tensile strength at room temperature and — 78° C, and arrested the fall in ductility resulting from the addition of 0·15% C. The cleavage strength at — 196° C was raised slightly, and the transition temperature in impact was lowered slightly. The properties after air cooling were thus improved only to a small extent by the introduction of 0·5% Mn.

In the furnace-cooled condition, the addition of manganese had similar effects in tension to those just mentioned for the air-cooled condition; these results are not reproduced here. It will be seen from Fig. 18 that the manganese resulted in a lowering of 30° C in the transition temperature in impact after furnace cooling compared with 5° C after air cooling.

Consideration of Results

Comparison of the effects on iron of phosphorus, carbon, and phosphorus plus carbon indicates that, although the properties of the ternary Fe–P–C alloys are determined to some extent by the phosphorus, they are governed mainly by the carbon. The influence of phosphorus in the ternary alloys is, of course, evident in the additional strengthening of the ferrite at normal temperatures. On the other hand, whereas phosphorus lowered the brittle strength of iron as determined at — 196° C by weakening the grain boundaries, the ternary alloys behaved much like Fe–C alloys in having strengths that were determined primarily by the carbon content, and that were on the whole little affected by the phosphorus content, with little sign of grain-boundary failure.

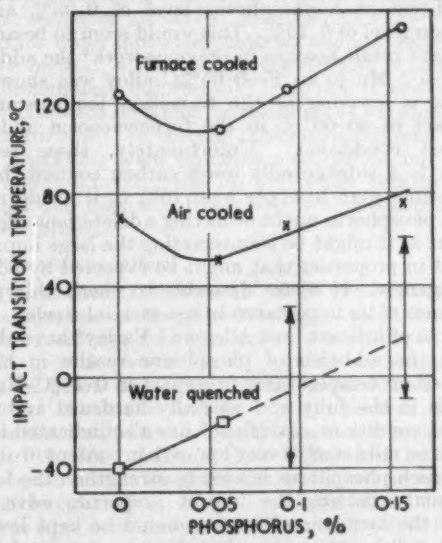


Fig. 15—Charpy impact transition temperatures of Fe–P–0·05%C alloys cooled at various rates from 950° C

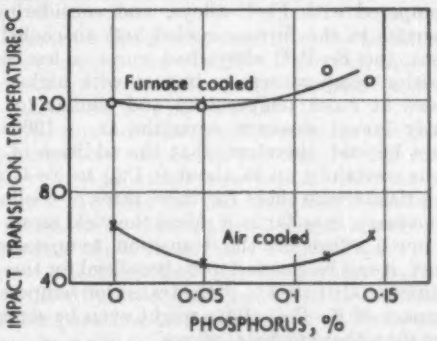


Fig. 16—Charpy impact transition temperatures of Fe–P–0·15%C alloys furnace-cooled and air-cooled from 950° C

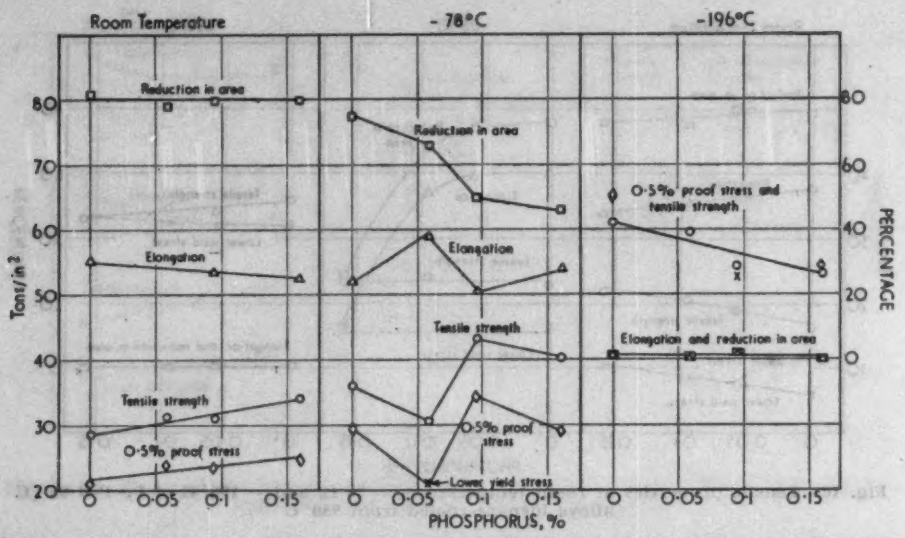


Fig. 12—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P–0·05%C alloys water-quenched from 950° C

Charpy Impact Properties

The temperatures of the transition from tough to brittle fracture in the V-notch Charpy impact test of the ternary Fe–P–C alloys after cooling at various rates from 950° C are shown in Figs. 15 and 16. The effect of phosphorus was comparatively small for the furnace-cooled and air-cooled conditions, but there was a general tendency for a minimum to appear in the curves at around 0·05% P. As this trend was not very pronounced, it is not surprising that it could not have been predicted from the yield-stress values at room temperature and the brittle strengths at — 196° C. The scatter in the results for the water-quenched condition at the higher phosphorus levels in the 0·05% C series has already been referred to in connection with the microstructures and tensile properties; nevertheless, the results indicated that for this condition there was a more consistent effect of phosphorus in raising the transition temperature. It will be recalled that there was a somewhat more pronounced drop in the brittle fracture stress at — 196° C of these alloys in the water-quenched condition.

The transition temperatures were lowered by increasing the rate of cooling. The difference between the transition temperatures of the furnace-cooled and water-quenched alloys in the 0·05% C series was not less than 120° C when the phosphorus contents were within commercial limits.

Comparison with Fig. 7 shows that in the furnace-cooled and air-cooled conditions the transition temperatures of the ternary Fe–P–C alloys were higher than those obtained in the binary Fe–P alloys. The fractures of the brittle specimens were similar to those

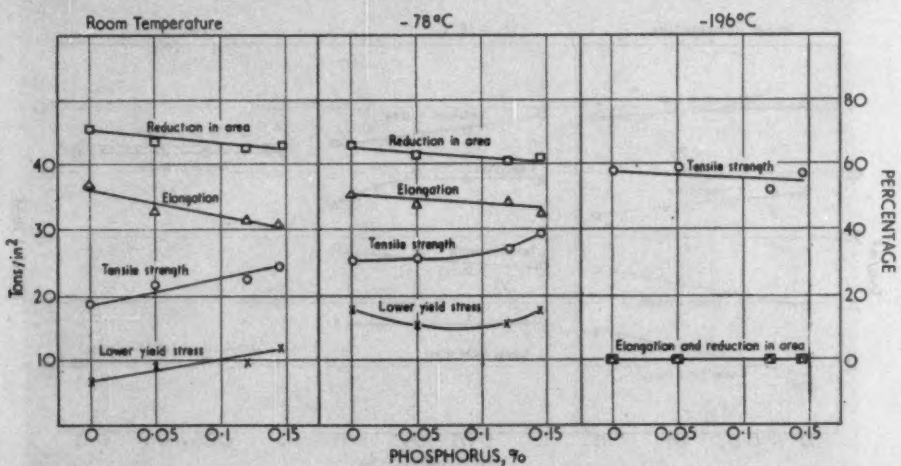


Fig. 13—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P–0·15%C alloys furnace-cooled from 950° C

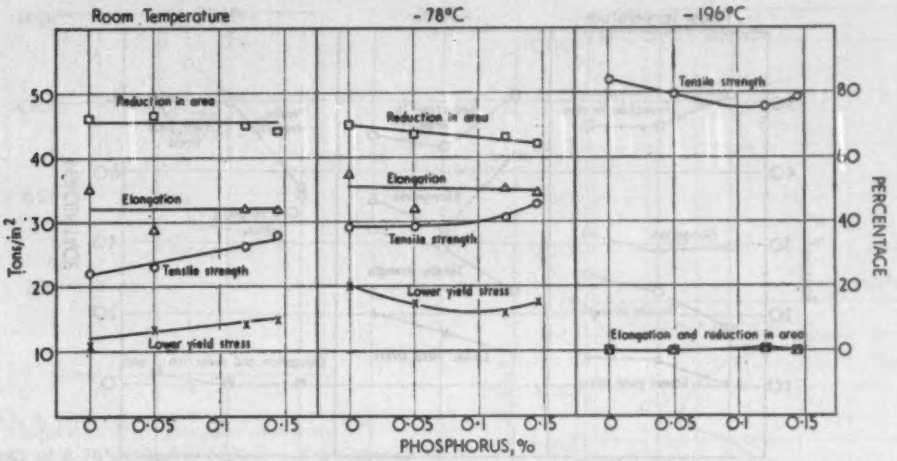


Fig. 14—Tensile properties at room temperature, — 78° C, and — 196° C of Fe–P–0·15%C alloys air-cooled from 950° C

of the tensile specimens tested at — 196° C in showing no very definite evidence of intergranular failure.

The Effect of Manganese

The alloy made to simulate a mild-steel composition by adding 0·5% Mn at a phosphorus content of 0·05% and a carbon content of 0·15% was tested in the air-cooled and furnace-cooled conditions. Its tensile properties and transition temperature in impact in the air-cooled condition are compared with those of two other alloys in Figs. 17 and 18. The manganese caused a small increase in yield stress and tensile strength at room temperature and — 78° C, and arrested the fall in ductility resulting from the addition of 0·15% C. The cleavage strength at — 196° C was raised slightly, and the transition temperature in impact was lowered slightly. The properties after air cooling were thus improved only to a small extent by the introduction of 0·5% Mn.

In the furnace-cooled condition, the addition of manganese had similar effects in tension to those just mentioned for the air-cooled condition; these results are not reproduced here. It will be seen from Fig. 18 that the manganese resulted in a lowering of 30° C in the transition temperature in impact after furnace cooling compared with 5° C after air cooling.

Consideration of Results

Comparison of the effects on iron of phosphorus, carbon, and phosphorus plus carbon indicates that, although the properties of the ternary Fe–P–C alloys are determined to some extent by the phosphorus, they are governed mainly by the carbon. The influence of phosphorus in the ternary alloys is, of course, evident in the additional strengthening of the ferrite at normal temperatures. On the other hand, whereas phosphorus lowered the brittle strength of iron as determined at — 196° C by weakening the grain boundaries, the ternary alloys behaved much like Fe–C alloys in having strengths that were determined primarily by the carbon content, and that were on the whole little affected by the phosphorus content, with little sign of grain-boundary failure.

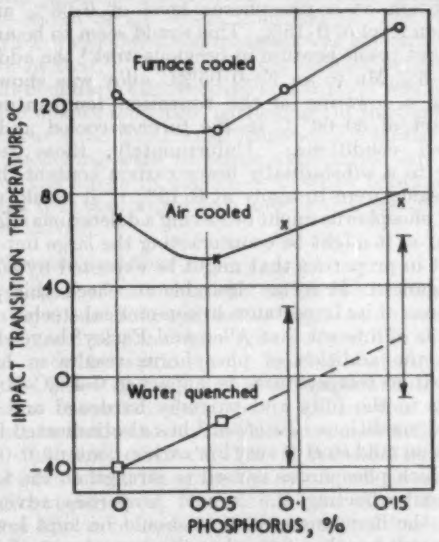


Fig. 15—Charpy impact transition temperatures of Fe–P–0·05%C alloys cooled at various rates from 950° C

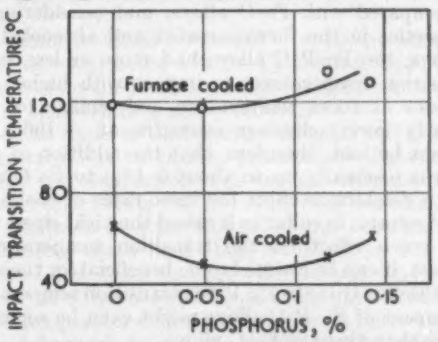
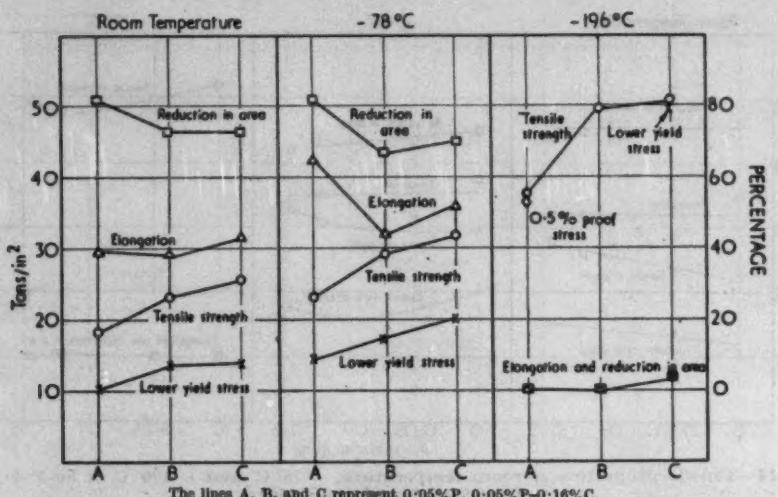


Fig. 16—Charpy impact transition temperatures of Fe–P–0·15%C alloys furnace-cooled and air-cooled from 950° C



The lines A, B, and C represent 0·05%P, 0·05%P–0·16%C, and 0·05%P–0·15%C–0·5%Mn alloys respectively

Fig. 17—Effect of 0·16%C and (0·15%C + 0·5%Mn) on the tensile properties at room temperature, — 78° C, and — 196° C of a 0·05%P alloy air-cooled from 950° C

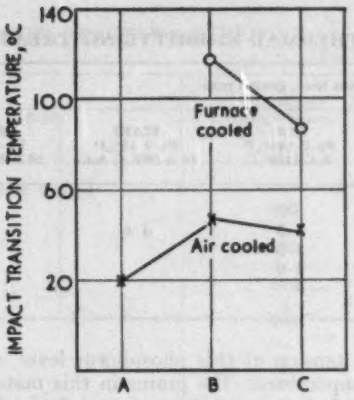
It has been seen in Part I that the deleterious effect of phosphorus in lowering the brittle strength of iron at — 196° C was reflected in the raising of the transition temperature in impact, which was no doubt also partly due to an enhanced yield-stress. In the furnace-cooled and air-cooled conditions, carbon caused some rise in the cleavage strength of iron at — 196° C but also caused a large rise in the transition temperature in impact; in the water-quenched condition, the effect of carbon in raising the cleavage strength of iron at — 196° C was accompanied by a considerable lowering of the transition temperature in impact. These factors have been discussed in detail in a previous publication.[4] For the furnace-cooled and air-cooled conditions, the transition temperatures in impact of the Fe–P–C alloys were higher than those of the Fe–P alloys for a given phosphorus content and heat-treatment in spite of the virtual absence of grain-boundary weakness in the ternary alloys. In the water-quenched condition, the influence of 0·05% C again predominated in giving impact transition temperatures in the ternary alloys that were generally much lower than in the Fe–P alloys.

Compared with Fe–C alloys, and considering the properties in the furnace-cooled and air-cooled conditions, the Fe–P–C alloys had more or less similar transition temperatures in impact with higher yield stresses at room temperature and similar (or even slightly lower) cleavage strengths at — 196° C. It cannot be said, therefore, that the addition of phosphorus (certainly up to about 0·1%) to Fe–C alloys had a deleterious effect for these rates of cooling; on the contrary, in so far as it raised the yield stress without much effect on the transition temperature in impact, it can be claimed to be beneficial for these two conditions. Up to 0·1% P, the transition temperatures in impact of Fe–P–C alloys might even be somewhat lower than those of Fe–C alloys.

Compared with Fe–C alloys, but considering the properties in the water-quenched condition, phosphorus can be said to be detrimental in the ternary alloys in lowering the brittle fracture stress at — 196° C and raising the transition temperature in impact. Nevertheless both the tensile properties and impact transition temperatures were improved by increasing the rate of cooling, and in a material containing 0·05% C and 0·05%P it was still possible to obtain a proof stress of over 20 tons/in² combined with a transition temperature of — 20° C in a Charpy **V**-notch impact test by water quenching.

The addition of 0·5% Mn has been shown to have only a slight effect in lowering the transition temperature in impact in the air-cooled condition, although it had a somewhat larger effect in the furnace-cooled condition at a phosphorus level of 0·05% and a carbon level of 0·15%. This would seem to be an important result because in previous work[4] the addition of 0·5% Mn to an Fe–0·05%C alloy was shown to cause a lowering of the transition temperature in impact of 50–60° C in the furnace-cooled and air-cooled conditions. Unfortunately, these results refer to a substantially lower carbon content, but if the same were to apply at 0·15% C, it would mean that phosphorus might be having a deleterious effect in so far as it might be counteracting the large improvement in properties that might be expected by adding manganese. It seems desirable to check this point because of its importance in commercial steels.

It is of interest that Allen and Earley[6] have shown that the addition of phosphorus results in higher transition temperatures in impact in 0·4%C–Mn–Mo steels in the fully and partially hardened and tempered conditions. Josefsson[7] has also indicated in his work on mild steel of very low carbon content (0·01%), in which phosphorus is used to strengthen the ferrite without affecting the impact properties adversely, that the manganese content should be kept low. It may well be, therefore, that the importance of phosphorus in mild steel lies mainly in its interaction with manganese.



The lines A, B, and C represent 0·05%P, 0·05%P–0·16%C, and 0·05%P–0·15%C–0·5%Mn alloys respectively

Fig. 18—Effect of 0·16%C and (0·15% C + 0·5%Mn) on the impact transition temperature of an Fe–0·05%P alloy heat-treated at 950° C

PART III—DEVELOPMENT OF GRAIN-BOUNDARY WEAKNESS BY ISOTHERMAL TREATMENT

It has been shown[2] that the grain-boundary weakness encountered in Fe–N alloys cooled at normal rates from the austenitic region can be developed in its most severe form by quenching from the temperature range 500–800° C approximately. Raising or lowering the temperature of treatment results in an improvement in grain-boundary strength as measured by the fracture stress at — 196° C, so that brittle fracture occurs to an increasing extent along the cleavage planes. The results of investigating grain-boundary weakness in Fe–P alloys by such isothermal treatments are given in this section.

It was concluded from the results in Parts I and II that carbon was effective in eliminating the grain-boundary weakness found in Fe–P alloys for material furnace cooled or air cooled from 950° C. The conclusion for water-quenched material was not quite so positive in that, although no very definite evidence of grain-boundary failure could be found, it was possible that a trace remained, particularly at the highest phosphorus level. Isothermal treatments followed by water quenching were therefore also carried out on some of the alloys in the Fe–P–C series because it appeared that they might be a more severe test of the effect of carbon.

Iron–Phosphorus Alloys

For the isothermal treatments on the two batches of Fe–P alloys, one of the difficulties encountered was to obtain a grain size that was suitable in so far as no further growth would occur during subsequent treatment below A_3, and that was comparable in all the alloys. It was also desirable not to heat-treat at a temperature inside the duplex field associated with the γ loop. A short time at 1150° C followed by air cooling gave a suitable grain size in the alloys made from Norwegian iron. For the alloys made from Japanese iron, a treatment of 16 h at 880° C followed by air cooling gave the most uniform grains but, even so, was not suitable for the lowest phosphorus content (0·044%) of this series in that there was a strong tendency to produce very coarse grains at temperatures just below A_3. Complete isothermal treatments were thus not carried out on this particular alloy.

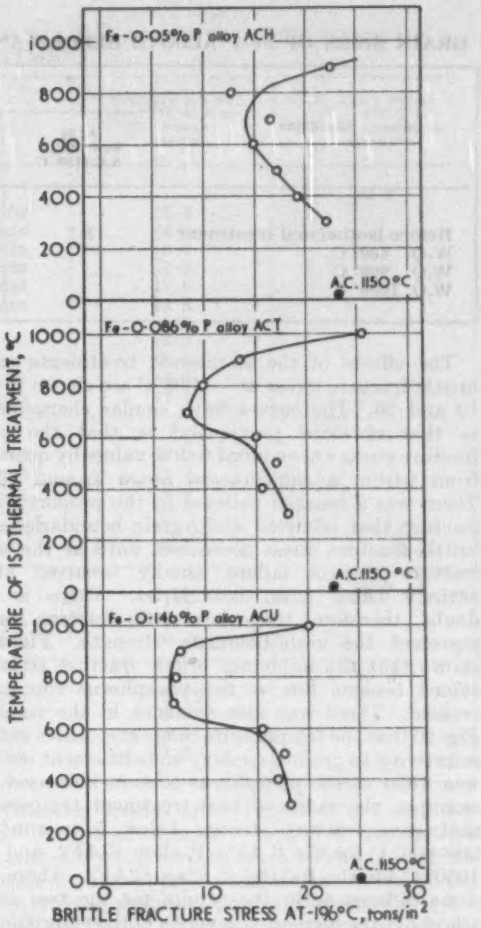


Fig. 19—Brittle fracture stresses at — 196° C of Fe–P alloys air-cooled from 1150° C and subsequently water-quenched from various temperatures. Alloys made from refined Norwegian electrolytic iron base

Measurements taken before and after the isothermal embrittling treatments (Table V) showed that the grain sizes remained substantially constant for treatments at temperatures below 1000° C.

The same times of isothermal treatment at various temperatures were adopted as previously.[2] The exact times are not reproduced here because they are not important apart from the fact that they were found to be adequate in that embrittlement occurred in much shorter times than those adopted, as will be seen later. The times had been worked out for Fe–N alloys so as to give equal opportunity for diffusion at each temperature, but adjustment of these times could not be made to suit the Fe–P alloys because no information seems to be available on the rate of diffusion of phosphorus in α-iron.

Table V

GRAIN SIZES OF Fe–P ALLOYS BEFORE AND AFTER ISOTHERMAL EMBRITTLING TREATMENTS

Condition	Grain Size, grains/mm				
	ACH Fe–0·05%P A.C.1150° C	ACT Fe–0·086%P A.C.1150° C	ACU Fe–0·146%P A.C.1150° C	87AF2 Fe–0·15%P 16 h 880° C A.C.	82AF2 Fe–0·31%P 16 h 880° C A.C.
Before isothermal treatment	3·7	3·2	3·5	4·6	4·5
W.Q. 800° C		3·4	3·8		
W.Q. 900° C		2·3	4·0		
W.Q. 1000° C		~7	~6		

The effects of the isothermal treatments on the brittle fracture stress at — 196° C are shown in Figs. 19 and 20. The curves have similar characteristics to that obtained previously[2] in that the brittle fracture stress was reduced to low values by quenching from within a temperature range around 700° C. There was a marked increase in the proportion of a fracture that occurred along grain boundaries as the brittle fracture stress decreased, until at the lowest fracture stresses failure usually occurred almost entirely along grain boundaries. There is little doubt, therefore, that the brittle fracture stresses represent the grain-boundary strength. Figure 19 shows that the minimum brittle fracture stress obtained became less as the phosphorus content increased. There was also evidence in the results in Fig. 20 that the temperature range over which extreme sensitivity to grain-boundary embrittlement occurred was wider as the phosphorus content increased. For example, the range of heat-treatment temperatures that gave fracture stresses below 5 tons/in² was 600–800° C for the 0·15% P alloy 87AF2, and 600–1050° C for the 0·31% P alloy 82AF2. There were some differences in the results for the two sets of alloys but the increasing severity of the embrittlement as the phosphorus content was raised was apparent in both series.

The times of heating required to develop the brittle fracture stresses recorded in Figs. 19 and 20 were investigated by heating the alloys for shorter times at selected temperatures. The results of these tests are given in Table VI. It will be seen that the time required to cause embrittlement is very short; the results indicate that the embrittling reaction was complete in 1 min. This was true for temperatures that gave only partial embrittlement as well as for those giving the minimum brittle fracture stresses.

The variation of brittle strength with temperature of testing of material embrittled in the temperature range developing the most severe grain-boundary weakness is shown in Fig. 21*a* and *b* for two alloys. The brittle strength of the Fe–0·31%P alloy 82AF2 increased with rise in testing temperature, but the material was still completely brittle at 100° C. The results given in Fig. 21*b* are for the Fe–0·044%P alloy 85AF1 in a very coarse-grained condition, and on which a full series of isothermal treatments was not carried out. The results for this alloy are of interest because they show the change from brittle to ductile failure, that is, the transition temperature range in tension at this phosphorus level was below room temperature. The grains in this material were so large that some of them occupied the full cross-section of the test-piece; at — 196° C the fracture occurred by separation of two grains at a stress of ¼ ton/in², whilst at 100° C failure occurred in one grain which gave a reduction in area of 100% with a chisel-edge fracture as would be expected of a single crystal.

The increase of intergranular strength with rise in testing temperature has been observed before[2] and is of considerable interest. In contrast, the cleavage strength of iron or steel has been found to fall as the testing temperature is raised, but the data in these cases are confined to the rather narrow temperature ranges of — 253° to — 196° C (iron single crystals[19]) and — 261° to — 211° C (0·2%C steel[20]).

In view of the possibility that the surface condition of the test-piece might affect the value of the fracture

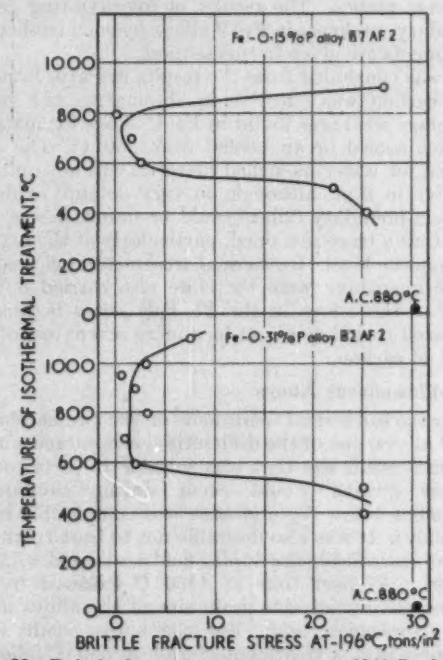


Fig. 20—Brittle fracture stresses at — 196° C of Fe–P alloys air-cooled from 880° C and subsequently water-quenched from various temperatures. Alloys made from refined Japanese electrolytic iron base

Table VI

BRITTLE FRACTURE STRESSES AT — 196° C OF Fe–P ALLOYS AFTER VARIOUS TIMES OF ISOTHERMAL TREATMENT

Alloy	Treatment Temperature, °C	Treatment Time	Fracture Stress at — 196° C after Treatment, tons/in²	Fracture Stress at — 196° C in Initial Condition, tons/in²
ACT Fe-0·086%P				22·4
	900	20 min	13·4	
	,,	1 min	14·0	
	700	1 h 36 min	8·4	
	,,	1 min	9·5	
	500	16 h 42 min	17·0	
	,,	1 min	18·5	
	400	94 h	15·5	
	,,	26 h	15·7	
	,,	2 h	17·4	
	,,	10 min	17·0	
	,,	1 min	17·4	
ACU Fe-0·146%P				24·0
	1000	1 h 45 min	20·0	
	,,	1 min	19·4	
	700	1 h 36 min	6·5	
	,,	1 min	8·8	
	500	16 h 42 min	17·5	
	,,	1 min	16·5	
87AF2 Fe-0·15%P				30·6
	700	2 h	1·5	
	,,	15 min	0·13	
	,,	1 min	0·25	
82AF2 Fe-0·31%P				30·1
	950	2 h 30 min	0·75	
	,,	30 min	1·0	
	,,	5 min	4·8	
	,,	1 min	1·5	

stress obtained at — 196° C for materials with very low strengths, two test-pieces of the Fe–0·146%P alloy ACU were embrittled at 700° C and the gauge length polished electrolytically. These test-pieces

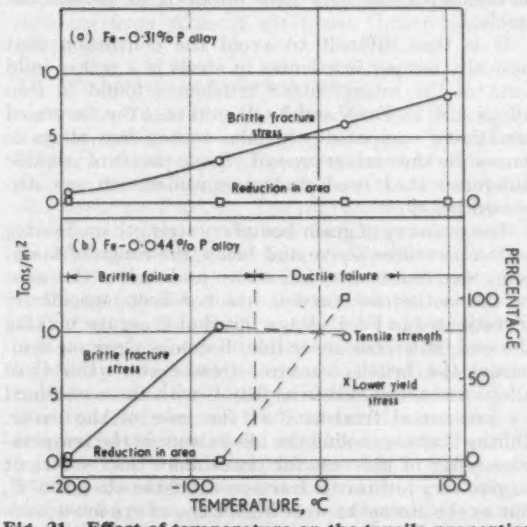


Fig. 21—Effect of temperature on the tensile properties of an Fe–0·31%P alloy and a coarse-grained Fe–0·044%P alloy after the following heat-treatment: air-cooled 880° C, water-quenched 700° C and 800° C respectively for the two alloys

gave fracture stresses at — 196° C of 6·8 and 6·6 tons/in² compared with 7·3 and 5·8 tons/in² for those with a normal surface finish. Thus the surface condition did not appear to affect the value of the fracture stress apart from a possibility that electrolytic polishing might improve the consistency of the results. Examination under the microscope of the surfaces of the electrolytically polished specimens after test revealed the presence of numerous twins, intergranular cracks, and slip, but the evidence for slip was confined to a few grains bounded by the fracture. For several reasons it is considered that the twinning and slip occurred during or subsequent to fracture along the grain boundaries.

Iron–Phosphorus–Carbon Alloys

Isothermal treatments were carried out on two alloys of the Fe–0·05%C series, one containing no phosphorus (ACM), and the other 0·16% (ACQ). In Table VII are given the results of treatments for various times at 500° C, preliminary experiments having indicated that embrittlement was faster in the phosphorus-containing alloy ACQ at this temperature than at 600° C. Treatment for times up to 48 h at 500° C gave a progressive but limited drop in brittle fracture stress at — 196° C for the phosphorus-containing alloy ACQ. That this was caused by grain-boundary weakening was shown by microscopic examination of sections through the fractures, from which it was seen that the reduction in fracture stress was accompanied by the development of partial

Fig. 22—Section through brittle fracture of Fe–0·15%P–0·05%C alloy *ACQ* after quenching from 500° C (48 h) showing grain-boundary failure ×500

failure along grain boundaries (Fig. 22). A longer time (120 h) at 500° C showed a recovery in fracture stress. The phosphorus-free alloy ACM showed no drop in brittle fracture stress after the same treatments, and the fractures in this alloy appeared to be entirely of the cleavage type.

An assessment of the extent of the embrittlement suffered by the phosphorus-containing alloy ACQ in terms of the rise in transition temperature in impact for a treatment of 48 h at 500° C showed that it was of considerable magnitude (Table VII). The degree of embrittlement was, nevertheless, insufficient to be reflected in the tensile properties at room temperature, which were as follows:

Upper yield stress	13·1 tons/in²
Lower yield stress	12·7 tons/in²
Tensile strength	24·5 tons/in²
Elongation	48%
Reduction in area	77%

These properties were similar to those obtained after air cooling from 950° C.

There are two striking differences between the sensitivities to embrittlement by isothermal heat-treatment of the Fe–P and Fe–P–C alloys. First, the degree of embrittlement obtainable in the Fe–P alloys as judged by the lowering of the brittle fracture stress at — 196° C was much more severe than in the Fe–P–C alloys. The embrittlement that could be induced in the Fe–P–C alloys, although comparatively mild, was nevertheless of the order encountered in the phenomenon of temper brittleness in low-alloy steels, when considered in terms of the rise in transition temperature in impact. Secondly, the time required to develop the brittleness in the Fe–P alloys, like the Fe–N alloys dealt with previously,[2] was very short indeed, 1 min at temperature being sufficient, whereas in the Fe–P–C alloys much longer times of the order of hours were required to develop the less severe degree of brittleness to which they were susceptible. Thus the addition of carbon to Fe–P alloys so reduced the severity and rate of development of the intergranular brittleness as to make it comparable with temper brittleness in commercial materials.

Comparison with Temper Brittleness

The phenomenon of intergranular brittleness in Fe–P and Fe–P–C alloys has characteristics similar to those encountered in temper brittleness in steels. The brittle fractures are partly intergranular as a result of a reduction of the grain-boundary strength below the cleavage strength in steels in the temper-brittle condition[8, 9] as in the alloys dealt with here. It has been established for Fe–P alloys that maximum embrittlement is obtained by heating in a certain temperature range; the embrittlement is less on heating at higher and lower temperatures. For steels there is also a temperature range for maximum embrittlement; for temperatures immediately above this range the degree of possible embrittlement becomes less, but for lower temperatures the times over which embrittlement occurs become so long that it has not been established how the degree of embrittlement attainable is related to the maximum embrittlement obtained at a higher temperature.[10] In Fe–P alloys the embrittlement occurs in a very short time, and the temperature range giving maximum embrittlement is somewhat higher than generally found in steels. In Fe–P–C alloys the rate of embrittlement and, apparently, the temperature range within which maximum embrittlement is obtained are both reduced to a level comparable with those encountered in steels. The embrittlement obtained in Fe–P alloys is also much more severe than in steels, but the addition of carbon again reduces the effect to a level comparable with that obtained in commercial steels.

It is thus difficult to avoid the conclusion that basically temper brittleness in steels is a rather mild form of the intergranular brittleness found in iron alloys such as Fe–N and Fe–P, and that the decreased sensitivity compared with the carbon-free alloys is caused by the carbon present. Some points of possible difference that need further consideration are discussed below.

The recovery of grain-boundary strength on heating at temperatures above and below the range of maximum embrittlement, i.e. above and below the nose of the isothermal curves, has not been specifically investigated in Fe–P alloys, but that it occurs, making the embrittlement reversible, becomes clear on comparing the brittle fracture stresses at — 196° C of alloys furnace cooled from 950° C with those obtained by isothermal treatment at the nose of the curve. During furnace cooling the alloys were in the temperature range of the nose for times more than sufficient to give very low brittle fracture strengths at — 196° C, but as the strengths were higher the alloys must have recovered during subsequent cooling from the nose. Likewise, rates of heating to temperatures above the nose that were sufficiently slow to cause severe embrittlement on passing through the nose gave

brittle fracture strengths at — 196° C characteristic of the temperature of treatment and cooling conditions. For example, specimens of the Fe–0·086%P alloy ACT when placed into a furnace at 1000° C, held for a given time and water quenched, gave a brittle fracture stress at — 196° C of 24·4 tons/in², whilst specimens treated similarly, except that they were heated up over a period of several hours with the furnace from cold, gave a value of 24·8 tons/in². With Fe–N alloys, it was shown[2] that furnace cooling from 950° to 625° C and water quenching gave the same brittle fracture strength at — 196° C as obtained on isothermal treatment at 600–700° C, whereas furnace cooling to room temperature gave much higher values. Recovery during isothermal treatment as distinct from slow cooling was shown to take place in the Fe–0·15%P–0·05%C alloy ACQ after prolonged heating at 500° C (Table VII).

Table VII

EFFECT OF PHOSPHORUS ON THE RESPONSE TO ISOTHERMAL TREATMENT AT 500° C OF AN Fe–0·05%C ALLOY

Condition	Fe–0·05%C, ACM	Fe–0·16%P–0·05%C, ACQ	
	Brittle Fracture Stress at — 196° C, tons/in²	Brittle Fracture Stress at — 196° C, tons/in²	Charpy Impact Transition Temperature, °C
A.C. 950° C	43·2	43·4	75
+ 1 h 500° C, W.Q.	44·8	...	...
+ 8 „ „ „	...	39·0	...
+ 16 „ „ „	44·2	35·3	...
+ 48 „ „ „	44·4	34·2	142
+ 120 „ „ „	43·8	41·2	...

For temper-brittleness in steels, recovery has been shown to occur on heating at temperatures above the nose.[9, 10] However, it does not seem to have been demonstrated whether it can occur at lower temperatures; the development of embrittlement in steels at low tempering temperatures becomes so slow that it is still proceeding at the rather long times that have been investigated, e.g. 240 h.[10] Gilbert[11] has shown that some recovery of temper embrittlement can occur during slow cooling of malleable and nodular cast irons.

In some steels at least, there is a further region just below Ac_1 in which embrittlement occurs, and which is separated from the usual range of temper brittleness ('lower nose') by a region of comparative freedom from embrittlement. This 'upper nose' embrittlement has been comparatively little investigated, and there is some doubt whether strictly it should be covered by the term temper-brittleness,[9] but the most recent work by Jaffe and Buffum[10] suggests that it should, although it has some characteristics that are different from those of the usual 'lower nose' embrittlement.

The Mechanism of Embrittlement

As with Fe–N alloys,[2] which behaved in a similar manner, no grain-boundary phase has been found either with the optical or electron microscopes to account for the low grain-boundary strengths obtainable in the Fe–P alloys. The explanation advanced for the shape of the isothermal embrittlement curve in Fe–N alloys would seem to be applicable to the Fe–P alloys. In the absence of evidence for a separate phase at the boundaries, this was based on the assumption that low grain-boundary strength was caused by segregation of the solute at the grain boundaries. At high temperatures, the solute atoms would be expected to be more or less randomly distributed in the material as a result of thermal fluctuations. On lowering the temperature, solute atoms would be expected to diffuse to the grain boundaries because this would involve less distortion of the matrix than if they remained in the interior of the grains. The grain-boundary strength would thus be progressively lowered until a maximum effect giving very low strengths were obtained. The subsequent increase of grain-boundary strength on further lowering the temperature to below the nose of the isothermal embrittlement curve is more difficult to explain, but a reasonable possibility is that clustering of the solute atoms at the grain boundary by diffusion along the boundary effectively denudes a large proportion of the boundary surface of solute atoms.

The segregation under appropriate conditions of solute atoms to grain boundaries and other regions of misfit has been suggested on numerous occasions. In 1938, Dean and Davey[12] claimed to demonstrate by chemical analysis the segregation of copper at the grain boundaries of a Zn–Cu alloy. Thomas and Chalmers[13] have claimed to have shown by an autoradiographic method the segregation of radioactive polonium in a Pb–5%Bi alloy, but an alternative interpretation of their experimental observations has been suggested recently.[14] McLean and Northcott[15] suggested grain-boundary segregation as the cause of temper-brittleness, although the details of their proposal involved embrittlement of the grain boundaries by the formation there of martensite resulting from carbon segregation. In the present work, carbon has the reverse effect of decreasing the grain-boundary weakness conferred by phosphorus.

Evidence has been obtained by Inman and Tipler, in work that arose from this investigation and which is reported separately,[16] that segration of phosphorus can occur at grain boundaries in iron. Phosphorus has a radioactive isotope ^{32}P which has a convenient half-life of 14 days. An alloy of 0·08% ^{32}P in iron was heat-treated so as to give brittle fractures that occurred almost entirely along grain boundaries. Small amounts of the order of 10^{-5} g were dissolved successively from the fracture surface, and the solutions analysed for phosphorus by means of a Geiger counter and for iron by a chemical method. For embrittled and unembrittled material the ratio of phosphorus to iron at the grain boundaries was found to be about 4 : 1.

By damping measurements, Lagerberg and Josefsson,[17] and Wert and Keefer,[18] have shown that the solubility of carbon is larger in fine-grained than in coarse-grained α-iron because of the greater concentration of carbon at the boundaries than in the interior of the grains. The influence of carbon in largely counteracting the embrittling effect of phosphorus could thus be caused by preferential segregation of carbon at the

boundaries, presumably by the carbon reaching the grain boundaries before the phosphorus and so interfering with its normal tendency to segregate there. On further holding at temperature it would be expected that some phosphorus would succeed in penetrating into the grain-boundary region, probably by displacing some of the carbon atoms.

It remains to be explained why phosphorus at the grain boundaries reduces their strength whereas carbon has no detectable effect. Any effect carbon alone might have on the strength of the grain boundaries would only be detected if they were made weaker than the cleavage strength of the crystals. The clue to this problem might lie in the temperature dependence of the intergranular strength. On cooling an alloy below the temperature range in which segregation occurred, an internal stress system could be developed at the grain boundaries if this were demanded by the relative coefficients of expansion of the crystal lattice and the phosphorus-laden grain-boundary regions. So long as they were always below the yield stress, such internal stresses would build up as the temperature was lowered, and so the observed grain-boundary strength would decrease progressively. An explanation along these lines assumes that internal stresses of such magnitude are not developed when the grain boundaries are filled with carbon atoms.

CONCLUSIONS

(1) Two important effects of adding phosphorus to iron are the much greater strengthening effect than obtained by comparable amounts of metallic alloying elements, and the progressive development of grain-boundary weakness which can be modified by heat treatment. For phosphorus contents above 0·05%, the transition temperatures in the Charpy impact test were high as a result of the enhanced yield stress and the reduction in brittle fracture stress arising from the weakening of the grain boundaries.

(2) In the ternary Fe–P–C alloys, the properties were determined mainly by the carbon content, and only in the water-quenched condition was there a progressive rise in transition temperature in impact as the phosphorus content was increased. For the slower rates of cooling, carbon eliminated the grain-boundary weakness of the binary Fe–P alloys; for the water-quenched condition, it was possible that a trace still remained.

(3) Attention is drawn to the good combination of tensile and impact properties obtained by the rapid cooling of material with a carbon content around 0·05%. The combination of properties was not quite as good as has previously been obtained with phosphorus-free materials. Nevertheless when the phosphorus contents are within commercial limits a material with very desirable properties can be obtained in this way.

(4) The rather small amount of work done to investigate the effect of adding manganese to an Fe–P–C alloy to simulate a mild-steel composition indicated that its beneficial influence was smaller than might have been expected. The possibility that the extent of the benefit to be derived from manganese in mild steel might be limited by the phosphorus content would appear to need further investigation.

(5) Isothermal treatments at various temperatures followed by water quenching showed that maximum intergranular weakness in the Fe–P alloys was obtained by heating in a certain temperature range; for temperatures above and below this range, the degree of embrittlement became less. The severity of the embrittlement, as judged by the minimum value of the brittle fracture stress at − 196° C, or by the range of temperatures of isothermal treatments that gave low values of brittle fracture stress, became greater with increase in phosphorus content. The time required to develop the degree of intergranular weakness appropriate to any temperature was very short indeed, times of the order of 1 min being sufficient.

(6) The presence of carbon very much reduced the severity of the intergranular weakness that could be produced by isothermal treatment, and also slowed down its rate of development. Nevertheless, the degree of intergranular weakness that could be induced in the Fe–P–C alloys, as given by the rise in transition temperature in impact, was of a similar magnitude to that encountered with temper brittleness in steels.

(7) Although recovery from intergranular embrittlement in the Fe–P alloys was not investigated in detail, it was clear that it could occur at temperatures above and below the nose of the isothermal embrittlement curves. In an Fe–P–C alloy, it was also shown that recovery could occur on holding for a long time at the temperature of embrittlement.

(8) In Fe–P alloys that had been severely embrittled by isothermal treatment, the brittle fracture stress (or intergranular strength) increased approximately linearly with rise in testing temperature. This is the reverse of the effect normally found for cleavage strength, although, in this case, data are confined to the much narrower temperature range − 261° to − 196° C.

(9) It is suggested that the results in general can be accounted for by grain-boundary segregation of phosphorus giving rise to a low intergranular strength, the influence of carbon being largely to prevent the segregation of phosphorus. Evidence is given in another paper that grain-boundary segregation of phosphorus can occur in iron. It is further suggested that segregation of phosphorus at grain boundaries might possibly cause internal stresses at lower temperatures, and so result in a progressive decrease in the observed value of the grain-boundary strength. For this explanation to hold, it would be necessary for internal stresses of such magnitude not to develop when carbon is the element segregating at the grain boundaries.

Acknowledgments

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The Isothermal Transformation of Spheroidized Pearlite to Austenite

By A. A. Golestaneh, C. Handford, and G. Bullock

SYNOPSIS

The isothermal transformation of spheroidized pearlite to austenite of a plain carbon eutectoid steel at supercritical temperatures has been determined using an electrical resistance technique. The influence of cementite distribution (number of spheroids per cm³) and transformation temperature on the rate of transformation has been examined. The incubation time appears to be a function of ferrite/cementite interfacial area, and for comparable values spheroidized pearlite has been found to begin transforming more quickly than lamellar pearlite. Above 740° C the transformation of lamellar pearlite is completed more quickly, however, while below this temperature the results are not sufficiently conclusive.

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Introduction

PREVIOUS WORK on the isothermal transformation of pearlite to austenite has been reviewed elsewhere[1] with the exception of the recent work by Molinder.[2] In his work the transformation to austenite of a spheroidized 1·27% carbon steel was followed by means of hardness, microscopic, and X-ray examinations. Results were obtained at different transformation temperatures (750–900° C) for only one initial structure but no attempt was made to determine the influence of initial structure on the transformation kinetics. The transformation was found to occur in three stages:

(i) Incubation period
(ii) Ferrite transforms to austenite with rapid dissolution of cementite
(iii) Remaining cementite dissolves in the austenite until equilibrium is achieved.

Molinder measured the volume of cementite, ferrite, and austenite in the specimen and the carbon content of the austenite as a function of transformation time. From these results the activation energy of stage (iii) was estimated as 200 kcal/mole.

As a continuation of earlier work on lamellar pearlite, the authors decided to examine the transformation of spheroidized pearlite to austenite using the same eutectoid steel and experimental technique to determine whether similar behaviour occurred. The eutectoid steel wire was heat-treated by either isothermal soaking at temperatures below A_1 or by 'cyclic annealing' between temperatures lying below and above A_1 in an effort to obtain uniformly spheroidized structures. This produced three spheroidized pearlite samples of somewhat differing mean cementite particle size, but in each sample there was a variation in particle size and distribution which probably influenced the time/transformation data to an unknown extent.

In previous work[1] the structure characteristic of lamellar pearlite has been defined by means of the interlamellar spacing. By analogy, values of the interspheroidal spacings of the spheroidized pearlites would appear to be desirable but it is difficult to envisage a method for calculating interspheroidal spacings which yields a significant result independent of the assumed distribution model. Therefore the structure characteristics of the spheroidized samples have been described by means of the number of cementite particles per unit volume as calculated from microsections. Mean number of cementite particles per cm² in the three samples are:

$$27{\cdot}5 \times 10^6,\ 19{\cdot}8 \times 10^6, \text{ and } 15{\cdot}2 \times 10^6$$

which, assuming random distribution, is equivalent to 144×10^9, $87{\cdot}8 \times 10^9$, and $59{\cdot}3 \times 10^9$/cm³ respectively. It has thus been possible, assuming uniform particle size and values for the density and compositions of ferrite and cementite, to calculate the interfacial ferrite/cementite grain-boundary areas

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The authors were members of the Department of Metallurgy at the College of Science and Technology, Manchester, when this work was carried out.

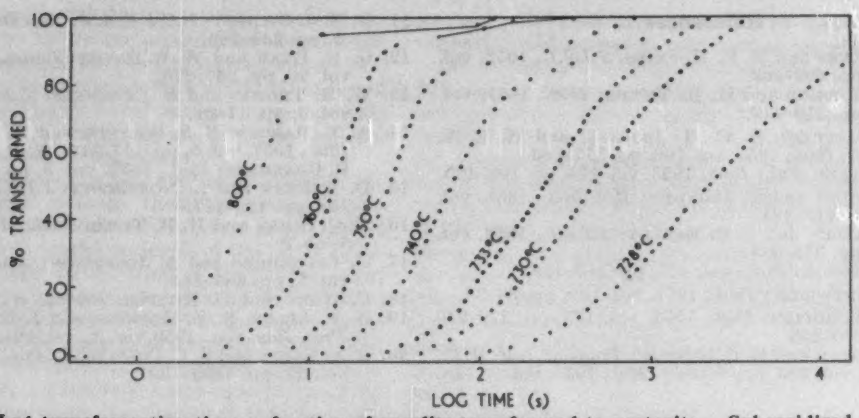

Fig. 1—Log transformation time *v.* fraction of pearlite transformed to austenite. Spheroidized pearlite has 144 × 10⁹ cementite spheroids per cm³

for each sample* and these are 6·4 × 10³, 5·4 × 10³, and 4·8 × 10³ cm²/cm³ respectively.

More accurate and representative values of ferrite/cementite grain boundary areas can be obtained if the size distribution of the cementite particles were known for each sample. To obtain such information, however, would appear to be a separate investigation. Therefore, for the present purpose the values obtained assuming uniform particle size have been used, but it must be remembered that their accuracy is not high.

The experimental method consists of heating the steel wire by means of a heavy alternating current to the desired isothermal transformation temperature in 3–5 s. A furnace which is already at the desired transformation temperature is moved over the specimen and the electrical resistance change in the latter, during transformation, is measured by the potential-fall method. Also the end of the transformation has been checked microscopically after quenching to martensite.

* Ferrite/cementite interfacial area = $1 \cdot 222n^{\frac{1}{3}}$ where n is the number of particles per cm³.

RESULTS AND DISCUSSION

Figures 1–3 show transformation/time curves whose shapes resemble those found for lamellar pearlites. The end of the reaction was considered to be completed when no cementite could be observed metallographically. These curves were expected to have the well-known sigmoidal shape associated with isothermal reactions which involve random nucleation. At the lower reaction temperatures this expectation appeared to be realized but at high temperatures (above 740° C) the reaction appeared to be more complex, as indicated by the presence of the 'knee' in the curves. This indicates the end of the stage in which ferrite transforms to austenite with rapid dissolution of cementite. The remainder of the transformation involves the dissolution of cementite in austenite and may also involve the partial removal of carbon-concentration gradients. Thus it would appear that the rate of reaction of ferrite and cementite is influenced by temperature to a much greater degree than the rate of solution of cementite in austenite in the temperature range considered in the present work.

Figure 4 shows values for the beginning and end of transformation as a function of temperature. In

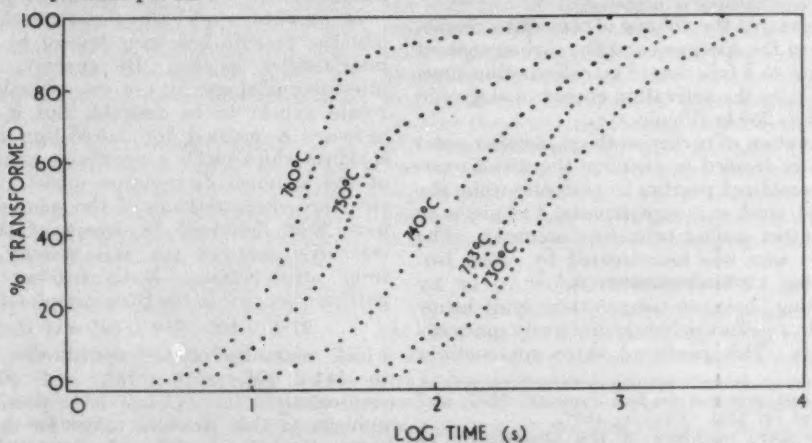

Fig. 2—Log transformation time *v.* fraction of pearlite transformed to austenite. Spheroidized pearlite has 87·8 × 10⁹ spheroids per cm³

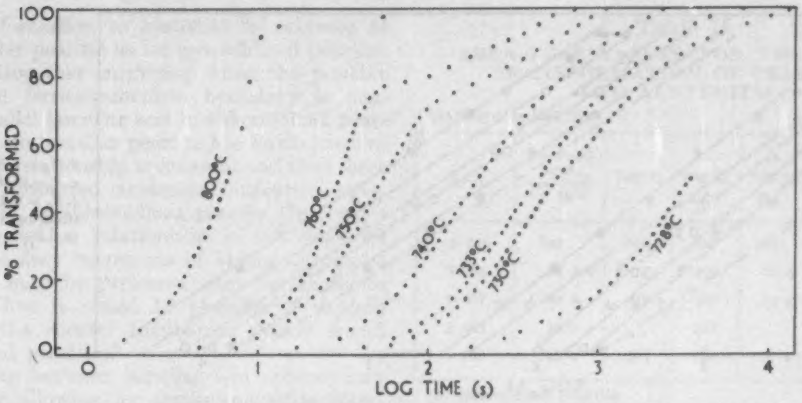

Fig. 3—Log transformation time *v.* fraction of pearlite transformed to austenite. Spheroidized pearlite has $59{\cdot}3 \times 10^9$ cementite spheroids per cm^3

previous work[1] the following equation had been found to apply for lamellar pearlites

$$\log t = A - \tfrac{3}{2} \log (\theta \times \theta_e) \quad . \quad . \quad . \quad (1)$$

where t is transformation time in seconds, A is a parameter, θ is the reaction temperature, and θ_e is the A_1 temperature for the steel (725° C).

The points in Fig. 4 show a significant departure from this relationship, especially at temperatures near A_1 and above 760° C where experimental conditions give greatest errors. Therefore there is little justification for determining numerical values for the parameter A in equation (1), which is a measure of the influence of the structure of the samples on the rate of reaction. This figure does show qualitatively, however, that the rate of transformation to austenite is more rapid for fine than for coarse spheroidized pearlite.

Figure 5 shows time values for the beginning and end of transformation as a function of temperature for lamellar pearlites. This figure, taken from earlier work,[1] enables lamellar and spheroidized pearlites to be compared.

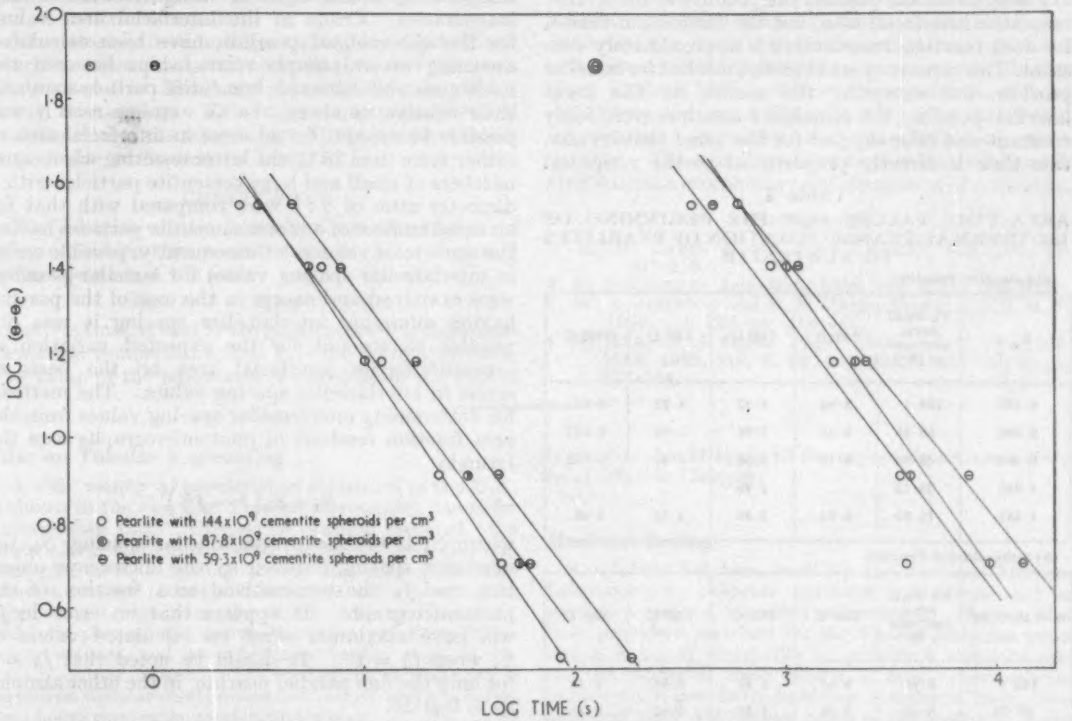
LOG (θ−θe)
LOG TIME (s)
○ Pearlite with 144x10⁹ cementite spheroids per cm³
◎ Pearlite with 87·8x10⁹ cementite spheroids per cm³
⊖ Pearlite with 59·3x10⁹ cementite spheroids per cm³

Fig. 4—Log transformation time *v.* log $(\theta - \theta_e)$ for beginning and end of isothermal transformation of spheroidized pearlite to austenite.

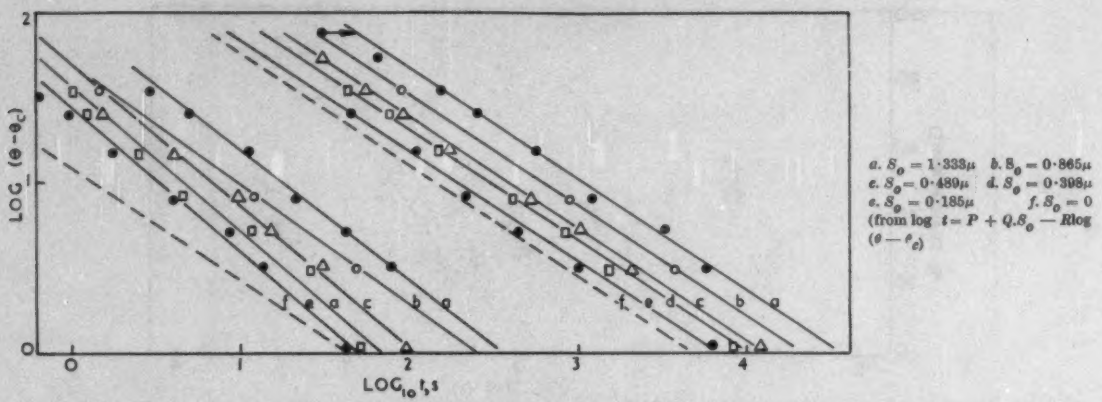

Fig. 5—Times of beginning and end of transformation shown as a function of reaction temperature. Groups of results on left and right correspond to beginning and end of transformation respectively; θ = reaction temp., θ_e = A_1 temp. of steel (725° C)

The time values for the beginning of transformation appear to be a function of ferrite cementite interfacial area. A selection of values for lamellar pearlite from previous work[1] and for spheroidized pearlites from present work is shown in Table I. The interfacial areas per cm³ for the lamellar pearlites have been calculated from the following equation

$$a = 10^4 \times 2/S_o$$

This assumes that the pearlite is ideally parallel and that the interlamellar spacing represents the distance between the centres of adjacent cementite lamellæ.

For spheroidized pearlite the product of the ferrite/cementite interfacial area and the incubation time t_b for each reaction temperature is approximately constant. This constancy is not so well marked for lamellar pearlite, but excepting the results for the finest lamellar pearlite, the remainder are, however, fairly constant and offer support for the belief that incubation time is directly proportional to the reciprocal of the ferrite/cementite interfacial area. It must be emphasized, however, that the area–time product is only approximately constant for lamellar and spheroidized pearlites when considered as separate groups. Comparison between the groups for a particular reaction temperature reveals that the area–time product is about twice as large for lamellar as for spheroidized pearlite. This difference may be due either to the rather crude models used when calculating interfacial areas or to inaccuracies in t_b values, the latter originating in lack of stability of the transformation temperature rather than in straightforward timing inaccuracies. Errors in the interfacial area values for the spheroidized pearlites have been calculated assuming several simple relationships between the maximum and minimum cementite particle-sizes and their relative numbers. In an extreme case it was possible to account for an error in interfacial area of rather more than 20%, the latter resulting when equal numbers of small and large cementite particles with a diameter ratio of 5 : 1 were compared with that for an equal number of uniform cementite particles having the same total volume. Consequently, possible errors in interlamellar spacing values for lamellar pearlites were examined and except in the case of the pearlite having minimum interlamellar spacing it was not possible to account for the expected variation in cementite/ferrite interfacial area on the basis of errors in interlamellar spacing values. The method[1] for determining interlamellar spacing values from the area fraction resolved in photomicrographs uses the formula

$$S_o = S_m\left(1 - f_N^{\frac{2}{3}}\right)^{\frac{1}{2}}$$

where S_o is the mean interlamellar spacing, S_m the minimum spacing resolved by the microscope objective, and f_N the non-resolved area fraction of the photomicrograph. It appears that an error in f_N will have maximum effect on calculated values of S_o when $f_N \simeq 1$. It should be noted that $f_N \simeq 1$ for only the fine parallel pearlite, in the other samples $f_N = 0$–$0{\cdot}25$.

Thus present results indicate that for a given ferrite/cementite interfacial area the beginning of the iso-

Table I

AREA–TIME VALUES FOR THE BEGINNING OF ISOTHERMAL TRANSFORMATION OF PEARLITES TO AUSTENITE

(a) Lamellar Pearlite)

S_o, μ	Fe–Fe_3C area, cm^2/cm^3 × 10^{-3}	730° C	733° C	740° C	750° C
0·185	108·1	8·96	4·32	1·72	0·81
0·398	50·25	5·42	2·98	1·15	0·552
0·489	40·90	6·16	3·06	1·23	0·532
0·865	23·12	...	2·95	...	...
1·333	15·00	6·82	3·30	1·32	0·60

(b) Spheroidized Pearlite

$n - 10^{-3}/cm^3$	Fe–Fe_3C area cm^2/cm^3 × 10^{-3}	730° C	733° C	740° C	750° C
143·9	6·41	3·22	1·57	0·61	0·29
87·78	5·43	3·26	1·59	0·62	0·29
59·26	4·77	3·24	1·60	0·62	0·30

thermal transformation to austenite takes twice as long for lamellar pearlite as for spheroidized pearlite. This is not altogether surprising when the possible nature of the ferrite/cementite boundary is considered in parallel lamellar and in spheroidized pearlites. Ideally, in lamellar pearlite the ferrite/cementite orientation relationship is constant and thus there would be no preferred austenite nucleation sites. Correspondingly, in spheroidized pearlite the ferrite/cementite orientation relationship is not constant and thus there may be regions of higher interfacial energy which may be preferred sites for austenite nucleation. Thus it would be possible to explain qualitatively the shorter incubation period found for spheroidized pearlites.

A comparison between lamellar and spheroidized pearlites, after allowing for ferrite/cementite interfacial area, of the time required to complete the transformation reveals little information (*see* Table II). The results for the spheroidized pearlite are not sufficiently consistent at the lower reaction temperatures. At higher reaction temperatures for comparable dispersion (on the basis of ferrite/cementite interfacial area) the transformation is completed more quickly for lamellar than spheroidized pearlites. This may be explained by the fact that the thickness of the cementite lamellæ lies between 200 and 1300 Å whereas the diameter of the spheroids lies between 10,000 and 50,000 Å. Since the final stages of the transformation involve solution of cementite in austenite, which is relatively slow,[2] the larger cementite particles in spheroidized pearlite may delay the end of the transformation.

The transformation/time data for spheroidized pearlite, when analysed according to the methods of Johnson and Mehl[3] and Avrami[4] did not yield a linear relationship. This suggests a more detailed examination of the morphology of the transformation is desirable to explain its kinetics.

CONCLUSION

The kinetics of the isothermal transformation of spheroidized pearlite to austenite can be represented by an equation of the form

$$\log t = A - B \log (\theta \times \theta_e)$$

where t is time, A and B are parameters, θ is the transformation temperature, and θ_e is the A_1 temperature. The value of the parameter A depends on the extent of transformation and the dispersion of the cementite spheroids. A greater degree of dispersion (i.e. a finer spheroidized sample) decreases its value. The parameter B is about 1·5, a similar value having been found previously for lamellar pearlite.

Table II

AREA–TIME VALUES FOR THE END OF TRANSFORMATION OF PEARLITES TO AUSTENITE

(*a*) Lamellar Pearlite

S_o, μ	Fe–Fe_3C area, $cm^2/cm^3 \times 10^{-3}$	730° C	733° C	740° C	750° C	760° C
0·185	108	603	229	118	47·9	...
0·398	50·3	417	200	72·4	38·0	21·0
0·489	40·9	417	209	72·4	36·3	22·0
0·865	23·1	...	219	...	...	23·0
1·333	15·0	501	186	87·1	39·8	23·2

(*b*) Spheroidized Pearlite

$n - 10^{-5}$/ cm^3	Fe–FeC_3 area, $cm^2/cm^3 \times 10^{-3}$	730° C	733° C	740° C	750° C	760° C
144	6·4	227	161	106	53·3	21·9
87·8	5·4	454	206	116	54·3	30·9
59·3	4·8	630	416	112	53·6	33·1

A comparison of transformation/time data for lamellar and spheroidized pearlites of similar dispersions, in the basis on ferrite/cementite interfacial area, reveals that the latter has the shorter incubation time. Above 740° C, where the results have been sufficiently consistent however, it appears that lamellar pearlite completes the transformation more quickly. These findings have been tentatively associated with the morphology and dispersion of cementite.

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Film on Tubular Engineering

A wide variety of tubular-steel structures in the U.K. is shown in the new film *Tubular Engineering*, made for Tubewrights Ltd. After some opening shots of 'the world's largest dragline' with its tubular-steel jib, and examples of floodlighting towers and masts, the film shows a number of less well-known structures such as a bow-string bridge over the River Aire in Yorkshire, which is part of the conveyor system of Ferrybridge power station, and an Admiralty testing tank of 12-million gallons capacity. The roof of the testing tank is carried on tubular-steel trusses and booms, and a tubular-steel boom carries the models under test.

The film is available through branch offices of the parent company, Stewarts and Lloyds Ltd.; overseas borrowers should apply to the export department of the head office in Glasgow.

Hardness Testing

A machine has been built by the National Physical Laboratory to calibrate hardness test blocks and so provide a national standard of hardness. The machine is to provide a standard for the Vickers hardness scale, but it is hoped eventually to provide a standard also for the Rockwell scale, which is used extensively in industry. When both hardness standards have been achieved, they will remove much of the confusion which exists today in hardness measurement throughout the world.

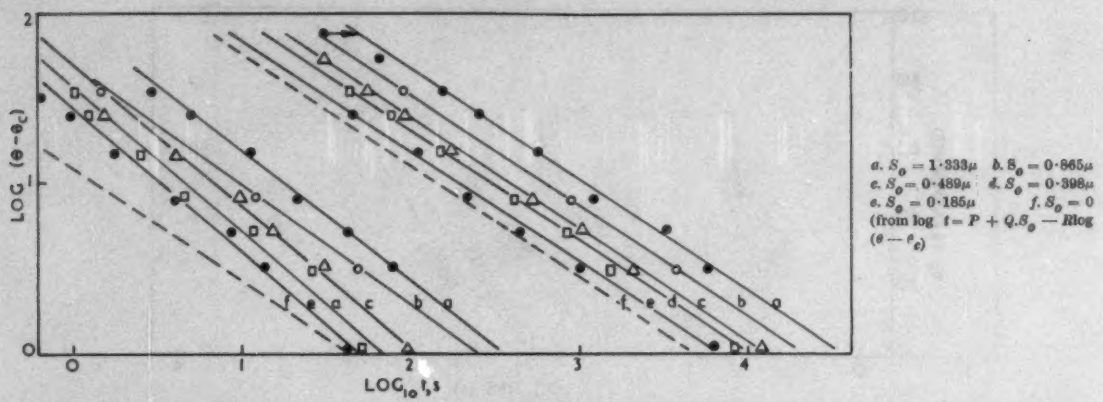

Fig. 5—Times of beginning and end of transformation shown as a function of reaction temperature. Groups of results on left and right correspond to beginning and end of transformation respectively; θ = reaction temp., $\theta_e = A_1$ temp. of steel (725° C)

The time values for the beginning of transformation appear to be a function of ferrite cementite interfacial area. A selection of values for lamellar pearlite from previous work[1] and for spheroidized pearlites from present work is shown in Table I. The interfacial areas per cm³ for the lamellar pearlites have been calculated from the following equation

$$a = 10^4 \times 2/S_o$$

This assumes that the pearlite is ideally parallel and that the interlamellar spacing represents the distance between the centres of adjacent cementite lamellæ.

For spheroidized pearlite the product of the ferrite/cementite interfacial area and the incubation time t_b for each reaction temperature is approximately constant. This constancy is not so well marked for lamellar pearlite, but excepting the results for the finest lamellar pearlite, the remainder are, however, fairly constant and offer support for the belief that incubation time is directly proportional to the reciprocal of the ferrite/cementite interfacial area. It must be emphasized, however, that the area–time product is only approximately constant for lamellar and spheroidized pearlites when considered as separate groups. Comparison between the groups for a particular reaction temperature reveals that the area–time product is about twice as large for lamellar as for spheroidized pearlite. This difference may be due either to the rather crude models used when calculating interfacial areas or to inaccuracies in t_b values, the latter originating in lack of stability of the transformation temperature rather than in straightforward timing inaccuracies. Errors in the interfacial area values for the spheroidized pearlites have been calculated assuming several simple relationships between the maximum and minimum cementite particle-sizes and their relative numbers. In an extreme case it was possible to account for an error in interfacial area of rather more than 20%, the latter resulting when equal numbers of small and large cementite particles with a diameter ratio of 5 : 1 were compared with that for an equal number of uniform cementite particles having the same total volume. Consequently, possible errors in interlamellar spacing values for lamellar pearlites were examined and except in the case of the pearlite having minimum interlamellar spacing it was not possible to account for the expected variation in cementite/ferrite interfacial area on the basis of errors in interlamellar spacing values. The method[1] for determining interlamellar spacing values from the area fraction resolved in photomicrographs uses the formula

$$S_o = S_m\left(1 - f_N^2\right)^{\frac{1}{2}}$$

where S_o is the mean interlamellar spacing, S_m the minimum spacing resolved by the microscope objective, and f_N the non-resolved area fraction of the photomicrograph. It appears that an error in f_N will have maximum effect on calculated values of S_o when $f_N \approx 1$. It should be noted that $f_N \approx 1$ for only the fine parallel pearlite, in the other samples $f_N = 0$–$0{\cdot}25$.

Thus present results indicate that for a given ferrite/cementite interfacial area the beginning of the iso-

Table I

AREA–TIME VALUES FOR THE BEGINNING OF ISOTHERMAL TRANSFORMATION OF PEARLITES TO AUSTENITE

(a) Lamellar Pearlite)

S_o, μ	Fe–Fe_3C area, cm²/cm³ × 10^{-3}	730° C	733° C	740° C	750° C
0·185	108·1	8·96	4·32	1·72	0·81
0·398	50·25	5·42	2·98	1·15	0·552
0·489	40·90	6·16	3·06	1·23	0·532
0·865	23·12	...	2·95	...	...
1·333	15·00	6·82	3·30	1·32	0·60

(b) Spheroidized Pearlite

$n - 10^{-5}$/cm³	Fe–Fe_3C area cm²/cm³ × 10^{-3}	730° C	733° C	740° C	750° C
143·9	6·41	3·22	1·57	0·61	0·29
87·78	5·43	3·26	1·59	0·62	0·29
59·26	4·77	3·24	1·60	0·62	0·30

thermal transformation to austenite takes twice as long for lamellar pearlite as for spheroidized pearlite. This is not altogether surprising when the possible nature of the ferrite/cementite boundary is considered in parallel lamellar and in spheroidized pearlites. Ideally, in lamellar pearlite the ferrite/cementite orientation relationship is constant and thus there would be no preferred austenite nucleation sites. Correspondingly, in spheroidized pearlite the ferrite/cementite orientation relationship is not constant and thus there may be regions of higher interfacial energy which may be preferred sites for austenite nucleation. Thus it would be possible to explain qualitatively the shorter incubation period found for spheroidized pearlites.

A comparison between lamellar and spheroidized pearlites, after allowing for ferrite/cementite interfacial area, of the time required to complete the transformation reveals little information (*see* Table II). The results for the spheroidized pearlite are not sufficiently consistent at the lower reaction temperatures. At higher reaction temperatures for comparable dispersion (on the basis of ferrite/cementite interfacial area) the transformation is completed more quickly for lamellar than spheroidized pearlites. This may be explained by the fact that the thickness of the cementite lamellæ lies between 200 and 1300 Å whereas the diameter of the spheroids lies between 10,000 and 50,000 Å. Since the final stages of the transformation involve solution of cementite in austenite, which is relatively slow,[2] the larger cementite particles in spheroidized pearlite may delay the end of the transformation.

The transformation/time data for spheroidized pearlite, when analysed according to the methods of Johnson and Mehl[3] and Avrami[4] did not yield a linear relationship. This suggests a more detailed examination of the morphology of the transformation is desirable to explain its kinetics.

CONCLUSION

The kinetics of the isothermal transformation of spheroidized pearlite to austenite can be represented by an equation of the form

$$\log t = A - B \log (\theta \times \theta_e)$$

where t is time, A and B are parameters, θ is the transformation temperature, and θ_e is the A_1 temperature. The value of the parameter A depends on the extent of transformation and the dispersion of the cementite spheroids. A greater degree of dispersion (i.e. a finer spheroidized sample) decreases its value. The parameter B is about 1·5, a similar value having been found previously for lamellar pearlite.

Table II

AREA–TIME VALUES FOR THE END OF TRANSFORMATION OF PEARLITES TO AUSTENITE

(*a*) Lamellar Pearlite

S_0, μ	Fe–Fe_3C area, $cm^2/cm^3 \times 10^{-3}$	730° C	733° C	740° C	750° C	760° C
0·185	108	683	229	118	47·9	...
0·398	50·3	417	200	72·4	38·0	21·0
0·489	40·9	417	209	72·4	36·3	22·0
0·865	23·1	...	219	...	...	23·0
1·333	15·0	501	186	87·1	39·8	23·2

(*b*) Spheroidized Pearlite

$n - 10^{-9}/cm^3$	Fe–FeC, area, $cm^2/cm^3 \times 10^{-3}$	730° C	733° C	740° C	750° C	760° C
144	6·4	227	161	106	53·2	21·9
87·8	5·4	484	206	116	54·3	30·9
59·3	4·8	630	416	112	53·6	33·1

A comparison of transformation/time data for lamellar and spheroidized pearlites of similar dispersions, in the basis on ferrite/cementite interfacial area, reveals that the latter has the shorter incubation time. Above 740° C, where the results have been sufficiently consistent however, it appears that lamellar pearlite completes the transformation more quickly. These findings have been tentatively associated with the morphology and dispersion of cementite.

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1. M. M. Labib *et al.*: *J.I.S.I.*, 1957, vol. 185, pp. 207–215.
2. G. Molinder: *Acta Met.*, 1956, vol. 4, pp. 565–571.
3. W. A. Johnson and R. F. Mehl: *Trans. A.I.M.M.E.*, 1939, vol. 135, pp. 416–441.
4. M. Avrami: *J. Chem. Phys.*, 1939, vol. 7, pp. 1103–1112; 1939, vol. 8, pp. 212–224; 1941, vol. 9, pp. 177–184.

Film on Tubular Engineering

A wide variety of tubular-steel structures in the U.K. is shown in the new film *Tubular Engineering*, made for Tubewrights Ltd. After some opening shots of 'the world's largest dragline' with its tubular-steel jib, and examples of floodlighting towers and masts, the film shows a number of less well-known structures such as a bow-string bridge over the River Aire in Yorkshire, which is part of the conveyor system of Ferrybridge power station, and an Admiralty testing tank of 12-million gallons capacity. The roof of the testing tank is carried on tubular-steel trusses and booms, and a tubular-steel boom carries the models under test.

The film is available through branch offices of the parent company, Stewarts and Lloyds Ltd.; overseas borrowers should apply to the export department of the head office in Glasgow.

Hardness Testing

A machine has been built by the National Physical Laboratory to calibrate hardness test blocks and so provide a national standard of hardness. The machine is to provide a standard for the Vickers hardness scale, but it is hoped eventually to provide a standard also for the Rockwell scale, which is used extensively in industry. When both hardness standards have been achieved, they will remove much of the confusion which exists today in hardness measurement throughout the world.

The Solubility of Nitrogen in Iron–Chromium Alloys

By E. T. Turkdogan and S. Ignatowicz

SYNOPSIS

The solubilities of nitrogen in iron–chromium alloys were determined at temperatures from 900° to 1400° C. In a few experiments the alloys were nitrided in ammonia–hydrogen mixtures at 600° C. X-ray examinations of the nitrided alloys and the extracted nitrides indicate that the nitride formed is CrN. The heat and entropy of solution of nitrogen in γ-iron–chromium alloys were derived from equilibrium measurements. The present results and the available data also permit the evaluation of the activity coefficient of nitrogen in α and liquid phases at 1400° and 1600° C. The effect of nitrogen on the austenite stability range has been shown. 1501

Introduction

THE RESULTS OF recent research on the behaviour of nitrogen in alloy steels indicate that this element has several beneficial effects. It has been shown by Westphal and Glatter[1] that alloy steels nitrided on their surfaces had wear resistance and antigalling characteristics when tested in high-temperature steam, i.e. up to about 250° C. Semkowicz[2] studied the physical properties of stainless steel containing various proportions of nitrogen. It was noted that some of the nickel could be replaced by nitrogen without impairing the properties of the steel. Tisinai, Stanley, and Samans[3] studied some constitutional properties of Fe–Cr–C–N stainless steels. They found that, in the presence of nitrogen and carbon, completely austenitic structures could be formed at temperatures above 1200° C in nickel-free stainless steels containing 21% to 33% Cr. Zackay, Carlson, and Jackson[4] prepared Cr–Mn austenitic steels by alloying them with nitrogen. These alloys had good mechanical properties in both the annealed and the cold-worked state.

Although the importance of nitrogen in stainless steel has been realized, no work has yet been done on any of the thermodynamic properties of nitrogen in these alloys, apart from some studies of the solubility of nitrogen in molten Fe–Cr alloys.[5, 6] It was therefore considered desirable to gain some knowledge on the limit of nitrogen solubility in Fe–Cr alloys.

EXPERIMENTAL DETAILS

The Fe–Cr alloys were equilibrated with 95% N_2 + 5% H_2 at various temperatures within the range 900–1400° C. In a few experiments, the alloys were nitrided in NH_3–H_2 mixtures at 600° C. The apparatus consisted of three horizontal furnaces which were provided with separate gas-purification trains and flow meters. The samples were introduced into the reaction chamber (1-in. bore recrystallized alumina) in a recrystallized alumina boat. Care was taken to avoid the oxidation of the sample during its introduction into and withdrawal from the reaction zone. For further experimental details, reference should be made to a paper by Corney and Turkdogan.[7]

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Dr. Turkdogan is Head of, and Mr. Ignatowicz is a member of, the Physical Chemistry Section of the Chemistry Department of B.I.S.R.A.

Materials

Three alloys were prepared from mixtures of pure iron and 99·9% Cr by melting in a high-frequency furnace. The iron used was pure B.I.S.R.A. AG iron, and the electrolytic chromium was purchased from Johnson Matthey and Company Limited. The melts were first purified in hydrogen for 30 min and then were put under vacuum for 30 min. To minimize the segregation of chromium, the melts were cooled under vacuum very rapidly through the liquidus–solidus range. This was accomplished easily in a high-frequency furnace without disturbing the crucible and its content. After solidification, the alloys were cooled very slowly to room temperature, and during the entire operation vacuum was retained. After removing the outer skin, fine turnings were obtained from the ingots, and were degreased. Representative samples were then analysed for chromium and nitrogen and the following compositions were obtained:

	Alloy 1	Alloy 2	Alloy 3
%Cr	4·76	8·67	14·10
%N	0·005	0·006	0·005

Chemical Extraction and X-ray Examination of the Nitrides

The nitrides in a few samples were extracted by a method similar to that used by Beeghly.[8] Some modifications, however, were made in the separation of the residue from the solution. The residue was in a very fine state of subdivision and separation by filtering was not possible, even when using very fine porous porcelain filters. The mixture was therefore centrifuged and the clear solution decanted. Several centrifugings and washings with methyl acetate were carried out to remove the traces of iron. The residue was then transferred to a weighed crucible, the methyl acetate evaporated under an infra-red lamp, and the crucible reweighed.

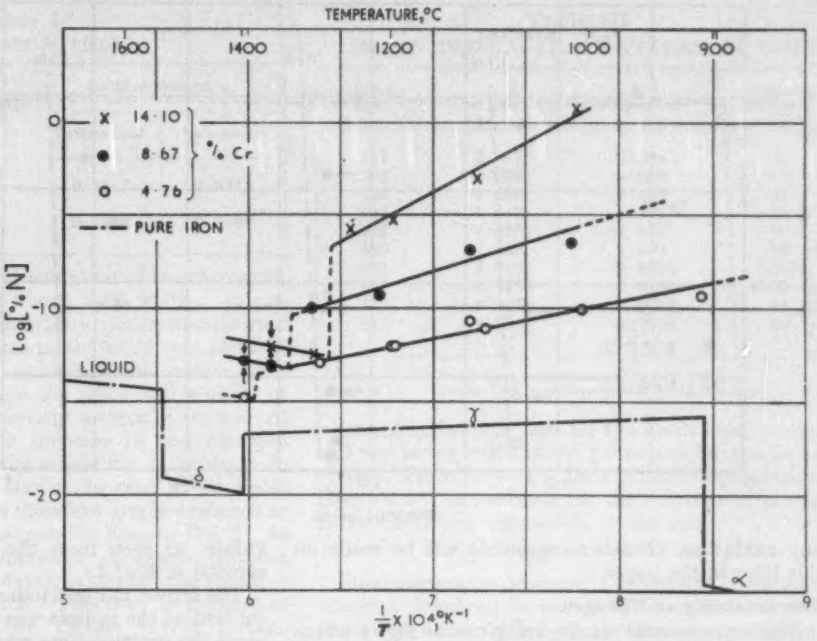

Fig. 1—Nitrogen solubility in Fe–Cr alloys at different temperatures under 1-atm pressure

X-ray powder photographs of the extracted nitrides and some of the nitrided alloys (prior to chemical treatment) were taken and the measured lattice spacings were compared with those of Cr_2N and CrN.

Analysis for Nitrogen

The solution of the sample in 1 : 1 hydrochloric acid, containing a few drops of hydrofluoric acid and hydrogen peroxide as recommended by Beeghly,[9] did not give satisfactory results with these alloys. Fuming in sulphuric acid, however, gave more consistent and reliable results, even with samples containing about 3·5% N which was present in the alloy as chromium nitride. The duplicate analyses agreed within ± 0·004%; this is considered quite satisfactory for the present purpose, because the alloys contained high percentages of nitrogen, e.g. from 0·03% to 3·5%.

RESULTS

Establishment of Equilibrium

In the previous work,[7, 10] using foils 0·002–0·006 in. thick, two to three days of reaction time was sufficient for the establishment of equilibrium between the alloy and nitrogen at 1 atm. In the present experiments about a gram of fine turnings were used and the time required to reach equilibrium was of the same order. Table I gives the results of a few experiments where equilibrium was approached from both sides.

In all subsequent experiments on nitrogen solubility at temperatures above 1100° C, the reaction time was from three to four days. Below 1100° C, however, the reaction time had to be increased to about one week for the establishment of equilibrium.

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			0·055	0·048
14·10	1360	2	0·0	0·064
			0·50	0·064

Oxidation of the Alloys

In the early stages of the work, attempts were made to measure the nitrogen solubility at temperatures below 900° C. Considerable difficulty was experienced in the solution of nitrogen in the chromium alloys at these moderate temperatures. When the samples were analysed for nitrogen, a considerable amount of green residue was found in the solution, even after repeated sulphuric acid fuming. X-ray examination indicated that the residue was Cr_2O_3. Oxidation was observed even when the alloys were heated in pure 'dried' hydrogen. Traces of water vapour could be responsible for this oxidation, because of the high affinity of chromium for oxygen. Extra columns of Anhydrone and a liquid-oxygen trap were included in the gas train, in an attempt to dry the gases still further. Even these special precautions did not prevent the oxidation of the samples and no further work was pursued at these temperatures.

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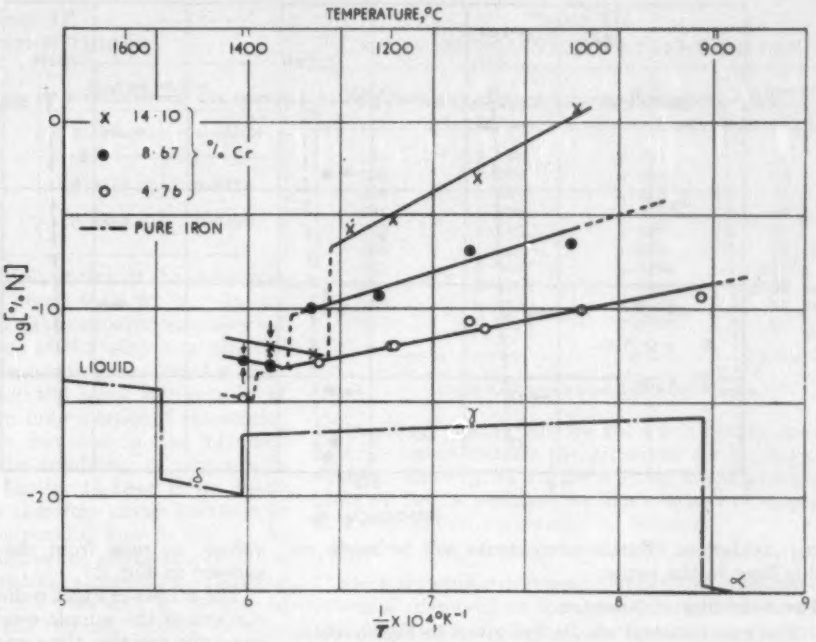

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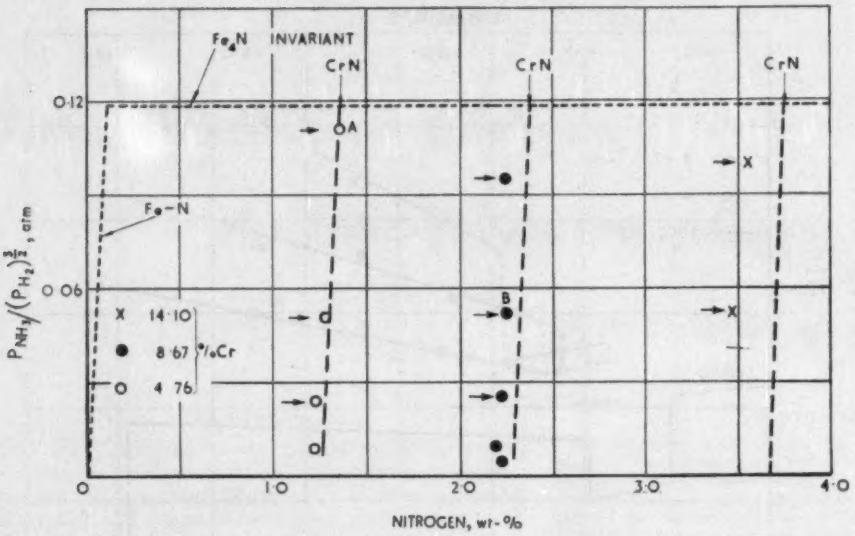

Parallel broken lines are calculated —*see* text

Fig. 2—Formation of chromium nitride in Fe–Cr alloys at 600° C

any oxidation. Further comments will be made on this later in the paper.

The Solubility of Nitrogen

The experimental results are given in Fig. 1 where log [N, %] is plotted against the reciprocal of the absolute temperature. If it is assumed that the solution of nitrogen in Fe–Cr alloys obeys Henry's law, for a given alloy and partial pressure of nitrogen (1 atm) the relationship in Fig. 1 would be linear. In fact the results in Fig. 1 support this view. Discontinuities of the lines indicate the phase transformation. The lower curve gives the change of nitrogen solubility with temperature in pure α, γ, δ, and liquid iron reproduced from a previous publication.[7] The arrows indicate the direction of approach to equilibrium (*see* Table I).

Nitride Formation

Three of the alloys were nitrided in NH_3–H_2 mixtures of known compositions at 600° C, and the results are plotted in Fig. 2 where the broken lines are drawn parallel to the solubility line for pure iron.[7] Although the traces of residual water vapour present in ammonia and hydrogen were not less than those present in the gases used for the nitrogen-solubility experiments, the alloys were not oxidized in NH_3–H_2 mixtures containing more than 4% NH_3. This may be attributed to the formation of a chromium nitride on the surface of the alloy when it is exposed to an atmosphere of high nitrogen potential. When the latter is reduced, the rate of nitride formation may not be fast enough to form the protective layer, and hence a trace of water vapour may oxidize the alloy. In fact, in experiments with alloys containing 4·76% and 8·67% Cr, the gas contained 1% or 0·5% NH_3, and after two weeks the alloys were found to be heavily tarnished and their nitrogen contents were very low, i.e. less than about 0·1%. And yet, when the previously nitrided alloys (nitrided in about 4% NH_3) were reheated for two weeks at 600° C in 1% or 0·5% NH_3, the samples were not tarnished and their nitrogen contents were very similar to the initial values, as seen from the points (not indicated by arrows) in Fig. 2.

The arrows in Fig. 2 indicate that the initial nitrogen content of the sample was very low. Although in all cases the reaction time was two weeks, the points for the higher chromium concentrations lie slightly on the left-hand side of the calculated lines. Further details will be given in the discussion.

In a few cases the nitrides were extracted very carefully and in samples *A* and *B*, marked in Fig. 2, the residue obtained was weighed. As seen from Table II the determined weights of residue agree very well with those calculated on the assumption that all the chromium in the alloy was converted to CrN.

The X-ray diffraction patterns obtained from a number of the extracts agree completely with one another and, as seen in Table III, the lattice spacings calculated from the powder photographs indicate that the extracted nitride is pure CrN, i.e. lattice constant $a = 4{\cdot}140$ Å as determined by Blix.[11] To ensure that the form of nitride was not affected by the chemical treatment during extraction, a few of the nitrided samples were ground and their X-ray powder photographs were taken. Measurements again indicated that the nitride formed was pure CrN.

DISCUSSION

Formation of CrN

On the above evidence, the reaction taking place in NH_3–H_2 mixtures may be written as follows:

$$CrN + \frac{3}{2}H_2 = Cr + NH_3 \quad \ldots\ldots\ldots\ldots\ldots\ldots(1)$$

For 600° C the equilibrium constant, K, of this reaction, derived from data compiled by Pearson and Ende,[12] is $2{\cdot}79 \times 10^{-6}$. Assuming that the Fe–Cr solid solutions obey Raoult's law (small deviations from ideality will have little or no effect on the present considerations) it is then possible to compute NH_3/H_2 ratios in equilibrium with CrN at any chromium concentration.

If CrN begins to form in an alloy at 600° C when $p_{NH_3}/(p_{H_2})^{\frac{3}{2}} = 0{\cdot}005$, i.e. about 0·5% NH_3 and the

Table II
EXTRACTED NITRIDES

Sample	Sample Weight, g	Residue Weight, g	
		Determined	Calculated
A	0·3542	0·0217	0·0213
B	0·3715	0·0394	0·0409

balance hydrogen, the concentration of the chromium in solid solution will be about 0·06 wt-%. It can therefore be stated that in all the experiments reported in Fig. 2 all the chromium in the alloy was nitrided completely at ammonia concentrations greater than about 1%. At this stage the alloy will consist of Fe–N solid solution (with only a trace of chromium) and CrN. Any further increase in the nitrogen potential will increase the solubility of nitrogen in the alloy in a manner similar to that in an Fe–N binary system. There is therefore ample evidence to justify the position of the parallel lines in Fig. 2. As the chromium content increases, difference between the determined and calculated values of total nitrogen becomes greater (*see* Fig. 2). This may be because in high-chromium alloys the nitride layers become denser which may retard the diffusion of nitrogen (and probably of chromium) through the nitride layer, before further reaction can take place.

Activity Coefficient of Nitrogen

In experiments where the solubility of nitrogen was determined, the reaction may be represented by

$$\tfrac{1}{2}N_2 = [N] \quad \ldots\ldots(2)$$

In pure iron this solution obeys Henry's law[7] and if it is assumed that the same is true for the solution of nitrogen in Fe–Cr alloys, then the following computations may be made. If the infinitely dilute solution is chosen as the standard for nitrogen, i.e. activity coefficient $f_N \rightarrow 1{\cdot}0$ when $\%\ N \rightarrow 0$, using weight percentages, the expressions in Table IV may be derived for the free energies of solution of nitrogen from the lines for γ-phase in Fig. 1.

The values for zero chromium are those given previously.[7] Uncertainties in the values of ΔH_N and ΔS_N for pure iron and 4·76% Cr alloy are less than $\pm$ 5%. For the other two high-chromium alloys, however, the following are the estimated limits of uncertainties: ΔH_N is within about $\pm$ 20% and ΔS_N is within about $\pm$ 15%. The marked effect of chromium on the heat and entropy of solution of nitrogen indicate strong attraction between these two atoms dissolved in iron.

The solubility of nitrogen in γ-Fe–Cr alloys can now be estimated for any pressure of nitrogen and any chromium concentration below 14·1% from the data in Table IV and the following expression:

$$\Delta H_N - T\Delta S_N = -4{\cdot}575\ T \log \frac{[\%\ N]}{\sqrt{p_{N_2}}} \quad \ldots\ldots(3)$$

where p_{N_2} is in atmospheres.

From the above values, the effect of chromium on the activity coefficient of nitrogen may be calculated as follows:

$$\Delta G_N - (\Delta G_N)' = 4{\cdot}575\ T \log \frac{f_N}{f_N'} \quad \ldots\ldots(4)$$

Table III
X-RAY DIFFRACTION PATTERN OF CrN

Index h k l	d, Å (Calculated)*	d, Å (Observed)†	Intensity (Visual)
111	2·391	2·394	S
200	2·070	2·068	VS
220	1·464	1·463	S
311	1·248	1·249	M
222	1·195	1·197	W
400	1·035	1·034	W
331	0·9497	0·9496	MW
420	0·9258	0·9260	M
422	0·8450	0·8460	M
511	0·7968	0·7979	W

* NaCl-type structure and $a = 4{\cdot}140$ Å
† Using CuK α-radiation and 9-cm camera

where $(\Delta G_N)'$ and f_N' are for the Fe–N binary system and the terms without the prime are for the ternary system. Since $f_N' = 1{\cdot}0$ for a given temperature and Henry's law is assumed for the solution of nitrogen in Fe–Cr alloys, expression (4) becomes

$$\Delta G_N - (\Delta G_N)' = 4{\cdot}575\ T \log f_N \quad \ldots\ldots(5)$$

Since chromium decreases the activity coefficient of nitrogen dissolved in iron, the nitrogen is expected to have a similar effect on the activity coefficient of chromium. This may be obtained by the application of the relationship derived by Wagner,[13] which in this case takes the form

$$\frac{\partial \log f_{Cr}}{\partial[\%\ N]} = \frac{52}{14} \times \frac{\partial \log f_N}{\partial[\%Cr]} \quad \ldots\ldots(6)$$

Application of equation (6) necessitates two assumptions: (*a*) the relationship derived by Wagner[13] for dilute solutions may be extended to the concentration range studied in this work and (*b*) the solution of chromium in iron obeys Henry's law. These are regarded as reasonable assumptions. Using the above expressions it is found, for example, that in a 10%Cr–0·17%N alloy the value of f_{Cr} at 1200° C is 0·88.

As seen from the solubility curves in Fig. 1, at temperatures above 1300° C the alloys are in the α-phase. The temperature range covered within this phase is not wide enough to determine the slopes of the lines which, as a first approximation, are drawn parallel to that for δ and α pure iron. The activity coefficient factor of nitrogen in the presence of chromium in α solid solution can now be derived, at least for 1400° C. The solubilities of nitrogen in liquid iron and Fe–Cr alloys were determined by several investigators.[5, 6, 14, 15] From these results the activity coefficient of nitrogen can again be derived for 1600° C. For comparison, in Table V derived values of f_N are given for the three chromium alloys in γ, α, and liquid phases.

It is interesting to note that although there are three phases between 1000° and 1600° C, for a given chromium concentration, the activity coefficient of nitrogen increases continuously with increasing temperature, except for the 4·76% Cr alloy at 1400° C. Therefore, it is assumed that the solubility value given in Fig. 1 for this alloy at 1400° C is high, i.e. it should have been about 0·023% N instead of 0·0338% as reported in Fig. 1.

Table IV

FREE ENERGIES OF SOLUTION OF NITROGEN IN IRON–CHROMIUM ALLOYS

(Infinitely dilute solution being the standard state for dissolved nitrogen)

% Cr	$\Delta G^{\circ}_{N} = \Delta H_{N} - T\Delta S_{N}$, cal/g-atom
0·00	$-2060 + 8{\cdot}9\,T$
4·76	$-9150 + 11{\cdot}7\,T$
8·67	$-13{,}800 + 13{\cdot}3\,T$
14·10	$-24{,}800 + 19{\cdot}5\,T$

Richardson and Dennis[16] studied some thermodynamic properties of Fe–Cr–C melts. From their results the carbon activity coefficient factors have been derived. For example, in molten iron containing 10% Cr and less than 0·3% C the value of f_C^{Cr} is about 0·6 which is higher than that of f_N^{Cr} (*see* Table V). This difference in values conforms with the free energies of formation of chromium carbide and nitride in liquid iron. The following values are derived by combining the available data:[12, 14, 15, 17]

$$Cr + [N] = CrN$$
$$\Delta G^{\circ}_{1873} = -3{\cdot}1 \text{ kcal/mean g-atom}$$
$$\tfrac{23}{6}Cr + [C] = \tfrac{1}{6}Cr_{23}C_6$$
$$\Delta G^{\circ}_{1873} = -1{\cdot}2 \text{ kcal/mean g-atom.}$$

In these calculations infinitely dilute solution was chosen as the standard state for nitrogen and carbon.

Effect of Nitrogen on γ-loop in Fe–Cr Alloys

The position of γ-loop in Fe–Cr solid solution is well established. This part of the phase diagram is reproduced in Fig. 3 from the existing data,[18] according to which, 14·1% Cr alloy should have been in the α-phase region, where the nitrogen solubility is much lower than that in the γ-phase. However, it is quite clear from the results in Fig. 1 that, in the presence of nitrogen from about 0·25% to 1·1% and within the temperature range about 1300–1000° C, this alloy is in the γ-phase. A sudden change in the nitrogen solubility shown in Fig. 1 for temperatures above about 1300° C indicates the phase transformation from γ to α-state. The points in Fig. 3 are the average $\gamma \rightarrow \alpha$ transformation values for different nitrogen concentrations and temperatures, derived from Fig. 1. Although the curves in Fig. 3 (broken lines) are only approximate, they indicate that small percentages of nitrogen extend appreciably the austenite composition range in Fe–Cr alloys.

CONCLUSIONS

(1) Chromium increases the solubility of nitrogen in γ, α, and liquid iron.

(2) The heat and entropy of solution of nitrogen in the γ-phase decreases with the increasing concentration of chromium. From these values, activity coefficients of nitrogen and chromium have been calculated.

(3) For a given chromium concentration, the activity coefficient of nitrogen increases with increasing temperature, although there may be phase transformations from γ to α and from α to liquid.

(4) The degree of ordering of chromium and nitrogen atoms dissolved in iron is much more pronounced than those of chromium and carbon atoms.

Table V

ACTIVITY COEFFICIENT OF NITROGEN DISSOLVED IN γ, α, OR LIQUID Fe–Cr ALLOYS

(Infinitely dilute solution being the standard state)

% Cr	f_N, γ-Phase		f_N, α-Phase	f_N, Liquid Phase*
	1000° C	1200° C	1400° C	1600° C
4·76	0·26	0·36	0·32	0·54
8·67	0·08	0·16	0·20	0·38
14·10	0·02	0·07	0·16	0·24

* Derived from available data[1, 6, 14, 15]

(5) The nitride formed at 600° C in the Fe–Cr alloys studied is CrN. This conclusion may also apply for other temperatures.

(6) The dissolved nitrogen increases the stability range of the γ-phase in Fe–Cr alloys. The present results indicate approximately the positions of the γ/α boundaries at 0·04%, 0·07%, and 0·10% N in the upper part of the γ-loop.

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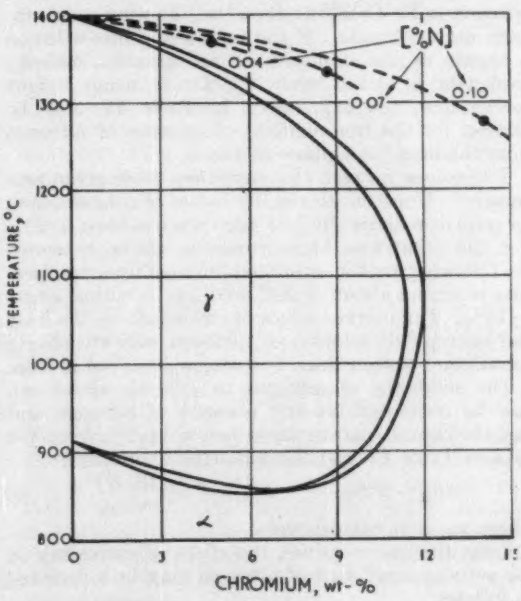

γ-loop in Fe–Cr from Kinzel and Crafts[18]

Fig. 3—Effect of dissolved nitrogen on the composition range of γ-loop in Fe–Cr alloys

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Veining in Ferrite

By A. Hultgren, Å. Josefsson, E. Kula, and G. Lagerberg

SYNOPSIS

Deformation structures in ferrite have been studied with the microscope, in particular veining, defined as sub-boundaries and obtained in the present work by hot deformation.

The sub-boundary network was found to show greater regularity of directions and to bemore closely meshed the higher the rate and the lower the temperature of deformation. A large grain size was also found to favour more regular sub-boundaries. In the same specimen three modes of etching of sub-boundaries could be observed, as (*a*) ridge veins, (*b*) groove veins, and (*c*) level veins, the last-named involving difference in relief between adjacent sub-grains. In any grain section these three modes may occur together under certain conditions, depending on orientation of section, carbon or nitrogen content, and rate of cooling.

Ridge veins were found to require certain minimum amounts of dissolved carbon or nitrogen, or both, for their appearance and to increase in amount with increase in these contents, carbon being more potent than nitrogen. Sub-boundaries in (100) sections always appear as groove veins. Groove veins in other sections are obliterated or transformed to low-relief ridges on slow cooling or ageing, probably as a result of precipitation. Level veining is interpreted as being caused by difference in orientation between adjacent sub-grains. 1434

IN THE PRESENT PAPER the terms 'veins' and 'veining' will be used for boundaries within the grains, visible after suitable etching; the terms are thus synonymous with sub-grain boundaries or sub-boundaries.

In an earlier investigation,[1] veins in ferrite were shown to form as a result of hot deformation in a temperature range below A_3. The veining pattern was influenced by temperature and rate of deformation, and the existing grain boundaries, after deformation, were often found to be modified by local movements so as to accommodate the veins, sometimes to the extent that grain boundaries and veins could not be readily distinguished. The veins appeared in the polished and etched section as ridges* or grooves. Recrystallized ferrite grains were found to be free from veining, thus confirming earlier observations. After mild steel had been cooled very slowly from 1000° C, no veining was found in the ferrite in some cases. After air cooling, however, some veining was generally present, presumably caused by strains accompanying the Ar_3 transformation.

Sub-boundaries in various metals have been studied subsequently and have been recognized as dislocation walls separating blocks of slightly differing orientation.[3–5] There seems no reason to doubt that veins in ferrite belong to that family of phenomena.

Samuel and Quarrell[6] limited their term 'α-veining' to sub-structures in ferrite produced during transformation $\gamma \rightarrow \alpha$ only and found by them to be particularly stable in the sense that they are unaffected by prolonged annealing in the α-range. Their findings will be discussed below.

In the present investigation ferrite veining resulting largely from hot deformation has been studied, with particular reference to the influence of different variables on the veining pattern and the type of veining relief.

EXPERIMENTAL PROCEDURE

Bending tests were made on 4 or 5 mm dia. × 25 mm rods, generally through an angle of 90° over a 5-mm radius, as in an earlier investigation,[1] at various temperatures and at slow or fast rates of bending (bending time: 11 s for slow and $\frac{1}{2}$ s for fast bending). The rods were heated in the preheated tongs for 3–5 min in a box furnace or in a lead bath, bent in the furnace, removed quickly, and generally quenched in brine or cooled in air. In certain tests wire specimens were deformed by pulling while in a

* Ridges along ferrite grain boundaries are not regarded as veins in the present investigation.[2]

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Table I

CHEMICAL COMPOSITION OF STEELS, wt-%

Steel	C	Si	Mn	P	S	Ti	N	Make
L	0·008		0·17	0·053	0·025		0·003	Domnarvet, high frequency
M	0·017		0·34	0·044	0·011		0·002	Domnarvet, high frequency
N	0·005		0·25	0·050	0·019		0·008	Domnarvet, high frequency
O	0·007–8		0·30	0·068	0·016		0·024	Domnarvet, high frequency
P	0·025		0·04	0·014		0·11	0·018	Domnarvet, high frequency
R	0·02		0·39	0·014	0·025	0·14	0·014	Domnarvet, high frequency
S	0·05	3·53	0·28	0·021	0·004		*	Surahammar, electric arc
X	0·003				0·002		max. 0·001	Electrolytic iron remelted *in vacuo*
K	0·02		0·08	0·014	0·010		0·002–3	Domnarvet, electric arc
T	0·020		0·10	0·008	0·016		0·009	Domnarvet, electric arc
LA	0·007†						nil	
LC	0·003						nil?	
LG	0·008						nil?	
LH	0·003						0·005	
KAK	max. 0·002						nil?	
KAA								
KBS								
KAN	max. 0·003						nil?	
KAG	0·005						nil?	
KAJ	0·003						0·025	
KBX								
KBR								
KBU	0·002						nil	
KBZ	0·003						0·003	
KCJ	0·004						0·008	
KCH	0·028						nil	
KCL								

* 0·32% Cu, 0·06% Ni

† gravimetric determination, probably too high

furnace. Bending tests were also made at room temperature followed by ageing at various temperatures.

A longitudinal section perpendicular to the bending axis was generally prepared for microscopical examination. The section was polished electrolytically[7] or mechanically and etched. To bring out any veining relief and sub-grain relief the etching reagents used were 4% picral followed by 1% nital; the 5-s etches were often repeated alternately three or five times. The micrographs shown were generally taken with oblique illumination, the incident light coming from the left.

Of particular interest was the bend region, where a scale of deformation existed ranging from maximum compression at the inside edge to maximum extension at the outside edge of the section. The observations recorded, where not otherwise stated, therefore refer to the heavily deformed areas of the bend. There was no essential difference between the deformation structures of the ferrite at the inside and outside edges, apart from the change in overall shape of the grains.

The chemical composition of the materials used and other data are given in Table I, and annealing treatments in Table II. The specimens, before being hot deformed, were generally normalized from 925° or 950° C and so already contained some veining. It was not considered necessary to remove this by recrystallization treatment because the veining pattern produced by hot deformation proved to depend on the conditions of deformation alone; i.e. the normalized material, on hot deformation, was found to give the same veining structures as a vein-free one of the same grain size so far as could be judged by microscopical observation. This was verified by results obtained on hot-deforming recrystallized material.

GENERAL REVIEW OF FERRITE VEINING AS AFFECTED BY DIFFERENT VARIABLES

Veins as observed in the microstructure of an etched section may differ in *relief* and in *pattern*. Classified by their *relief*, they may be:

(i) Ridge veins (Figs. 17*c* and *e*)
(ii) Groove veins (Fig. 19*c*)
(iii) Level veins (Figs. 3 and 8), the latter forming boundaries between blocks of different level.

In the same grain section, these three types may be combined in various ways.

The veining *pattern* may be openly or closely spaced (Figs. 18*a* and 19*a*), may form an irregular network (Fig. 17*e*), or may consist of two or three sets of intersecting parallel lines, more or less straight (Figs. 1 and 19*a*). In level veining, the sub-grains may form alternating levels (Fig. 1) or a series of steps (Fig. 2). In a polycrystalline specimen of ferrite, particularly after etching in nital, some grain sections may be nearly passive in the sense that the matrix appears to have been attacked very slightly or not at all (Figs. 9 and 13). This is an orientation effect and will be discussed below. Furthermore, as will be shown, the same sub-boundaries may appear as different types of veins when sectioned in different planes, e.g. as level veining in one plane and as groove veining in another.

The main variables that affect the ferrite veining pattern produced by hot deformation are grain size,

Table II
ANNEALING TREATMENTS

Steel	Starting Condition	Treatment No.	Temperature, °C	Atmosphere	Time, h
LA	*L*		690–700	H_2 + 30% H_2O	48
LC	*L*		700	H_2 + 20–25% H_2O	168
LG	*LC*		700	H_2 passed over naphthalene at 0° C	120
LH	*LC*		580	H_2 bubbled through aqueous NH_3 solution (100 cm³ NH_3*/1000 cm³) at room temperature	120
KAK	*K*	1	1200	H_2†	357
KAA		2	950	H_2‡	240
		3	700	H_2‡	120
KAN	*K*	1	950	H_2 bubbled through distilled water at room temperature	360
		2	950	H_2‡	240
		3	700	H_2‡	120
KAG	*K*	1	1200	H_2†	357
		2	950	H_2‡	240
		3	700	H_2‡	120
		4	650	H_2‡ bubbled through *n*-heptane at room temperature	72
KAJ	*K*	1	950	H_2 bubbled through distilled water at room temperature	360
		2	950	H_2‡	240
		3	700	H_2‡	120
		4	580	H_2§ bubbled through aqueous NH_3 solution (150 cm³ NH_3*)/1000 cm³) at room temperature	72
KBS	*KAK*	1	950	H_2‡	1
		2	700	H_2‡	1
KBR	*KAK*		580	H_2§ bubbled through aqueous NH_3 solution (150 cm³ NH_3*/1000 cm³ at room temperature	3
KBX					
KBU	*K*		1200	H_2 bubbled through distilled water at room temperature. Cooled in argon	336
KBZ	*KBU*		580	H_2‖ bubbled through aqueous NH_3 solution (40 cm³ NH_3*/1000 cm³) at room temperature	96
KCJ	*KBU*		580	H_2‖ bubbled through aqueous NH_3 solution (80 cm³ NH_3*/1000 cm³) at room temperature	96
KCH	*KBU*		710	H_2 bubbled through *n*-heptane at room temperature	24
KCL					

* sp. gr. 0·91. †purified: P_4O_{10} ‡purified: H_2SO_4, P O_5, 40% FeTi (870° C, pea size); § purified: H_2SO_4; ‖ purified: KOH.

temperature, and rate and degree of deformation, whereas the type of veining relief is influenced by orientation of section, rate of cooling after deformation and the amounts of dissolved carbon or nitrogen or both.

Ridges along grain boundaries (Figs. 18*b* and *c*) and the ridge veins studied in this investigation show apparent similarity, in that both have a rounded relief. It was early suspected that ridge veining resulted from dissolved carbon or nitrogen, or both, and the absence of ridge veining in hot-deformed ferrite was finally achieved only after numerous refining treatments for decreasing the content of those elements had been tried.

If recrystallization occurs during or immediately after deformation some vein-free grains will generally be found. For steel *M* it was found that for slow bending no recrystallization occurred up to 750° C, whereas for rapid bending recrystallization started on exceeding 730° C.

The effect of separate variables will be reported under different headings, including also the results of cold-deformation tests and ageing and reheating tests.

Degree of Deformation

A specimen of steel *M* (0·017% C, 0·002% N), normalized (grain size A.S.T.M. 7), was bent rapidly at 850° C and quenched. Figure 17*a* shows the structure of the middle region, Figs. 17*b* and 17*c* that of the stretched region, and Figs. 17*d* and 17*e* that of the upset region of the bend. In Fig. 17*a* the grain boundaries have remained smooth, no recrystallization has occurred, and little veining is to be seen. In the stretched and upset regions, however, the deformed grains show profuse veining; these are mainly ridge veins forming a network of high and lower relief, but there is also some level veining, as in the grain to the lower right in Fig. 17*e*. In these regions grain boundaries are irregular and modified. Some vein-free recrystallized grains are also present forming smooth boundaries with one another (Figs. 17*b*, *d*, and *e*). The amount of veining obviously increased with degree of deformation. The dark areas are transformed austenite.

The effect of increasing deformation (slow) at 750° C was studied on normalized specimens of steel *M*. In one series, the bending angle was varied from 18° to 90°; in another, the specimens were bent successively in stages of 18°, with an interval of 30 s between the stages. All the specimens were quenched after bending. The type of structures obtained was essentially similar for the two series; i.e. it was determined in each case by the amount of deformation alone. With increasing bending angle, the amount of veining, both ridge and groove, and the irregularity of grain boundaries increased.

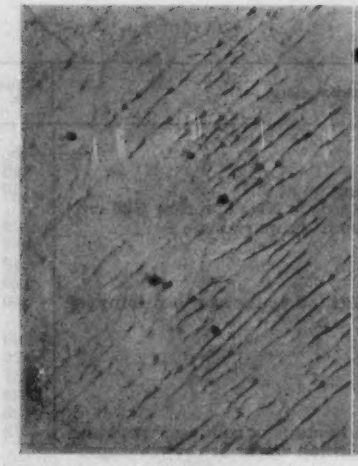


Fig. 1 — Electrolytic iron *X*, normalized, bent rapidly at 730° C and quenched; bands of alternating level, ridge veins × 1200

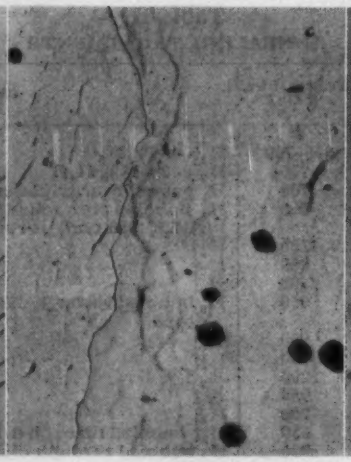


Fig. 2—Steel *LA*, bent slowly at 750° C and air cooled; grain-boundary ridge, ridge veins and irregular level veins × 1200

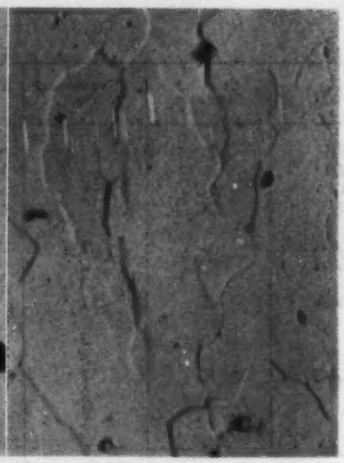


Fig. 3—Steel *LA*, same specimen; irregular level veins × 1200

Rate of Deformation

For steel *M*, the effects of slow and rapid deformation at 750° C are shown in Figs. 18*a* and 19*a* respectively. In both cases ridge and groove veins intersect. In the latter case, the veining pattern is more closely spaced and regular than in the former. All sub-grains appear to be of the same level. Figure 19*c* shows a variant with intersecting sets of groove veins.

For electrolytic iron *X*, which after normalizing at 950° C was very coarse-grained (A.S.T.M. 0), a similar comparison was made, with similar results. In addition, after rapid bending, level veining occurred, with intersecting bands of alternating level, sometimes separated by ridge veins (Fig. 1).

Steel *LA*, of low carbon content (grain size A.S.T.M. 7), showed similar structures after such treatments, with irregular level veining occurring after slow bending (Figs. 2 and 3), and banded level veining after rapid bending. Some ridge veining was also present (Fig. 2).

Temperature of Deformation

Specimens of steel *M* were slowly bent at 750°, 625°, 600°, 550°, and 500° C, and cooled in air. As illustrated in Figs. 4 (750° C), 5 (600° C), and 6 (550° C), on lowering the temperature of deformation, the rate being kept constant, the veining pattern became finer and showed greater regularity of direction and the grain boundaries became less disturbed. Deformation at room temperature produces no visible veins in the ferrite, but the structure obtained, showing intersecting bands of rounded relief (Fig. 7*a*), is of interest in this connection and will be referred to later.

Holding at Temperature after Deformation

A series of specimens of steel *LA* were bent slowly at 750° C, held at this temperature for 0–10 min, and quenched. In all specimens there was a moderate amount of ridge veining present, seemingly independent of holding time within the times given.

Grain Size

After a grain-coarsening treatment at 1000° C, the veining pattern produced by hot-bending at 750° C in steel *M* (grain size A.S.T.M. 5–6) showed a tendency towards more regularity of directions than in the normalized condition (grain size A.S.T.M. 7).

Rate of Cooling after Deformation

Figure 18*a* shows the structure of a specimen of steel *M* slowly bent at 750° C and quenched; Fig. 20 shows that of a similar specimen slowly bent at 700° C and then slowly furnace-cooled. The veining pattern is similar in the two photomicrographs and ridge veins of high relief are present in both, but the groove veins of the former are replaced by ridges of low relief in the latter. Similar effects of ageing will be discussed below.

In comparative tests with specimens of steels *L*, *M*, *N*, and *O*, bent slowly and quenched, it was found that steel *L* was similar to *M* in regard to presence of ridge and groove veins, whereas steels *N* and *O* showed ridge veins of high and low relief but no groove veins, even after quenching.

It should be emphasized that the greater number of etching treatments given on p. 248 (up to 5 times) was generally required to bring out the groove veins. A more detailed study of etching variables might have revealed conditions for bringing out groove veining in steels *N* and *O* as well.

To investigate whether ridge veins were affected by the method of cooling, 1·6-mm thick discs were held for 5 min at 700° C and then cooled in different ways. One was quenched in 10% NaOH solution, another rapidly transferred to a lead bath at 600° C, held 5 min, and quenched, and a third was treated similarly at 550° C. Ridge veins of similar appearance were present in all specimens, to a small but similar extent.

Composition

The effects of hydrogen, silicon, and titanium, and

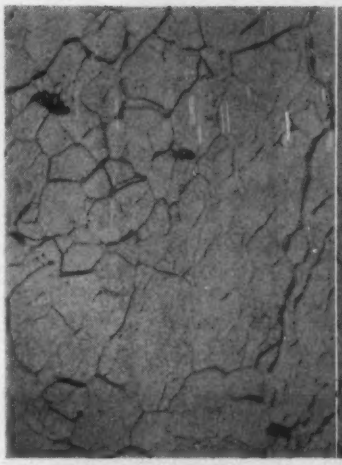


Fig. 4—Steel *M*, normalized, bent slowly at 750° C, and air cooled; ragged grain boundaries, irregular ridge veining of high and low relief × 1200

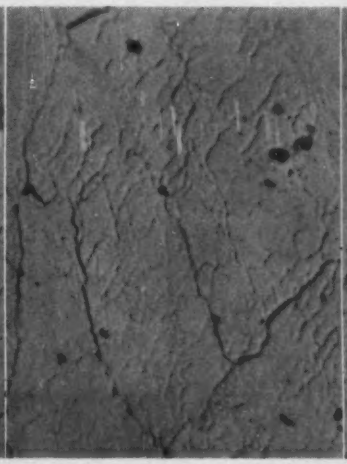


Fig. 5—Steel *M*, normalized, bent slowly at 600° C and air cooled; less ragged grain boundaries, ridge veins of closer spacing × 1200

Fig. 6—Steel *M*, normalized, bent slowly at 550° C and air cooled; fairly smooth grain boundaries, sets of closely spaced parallel ridge veins × 1200

of lowered carbon and nitrogen contents on ridge veining were studied.

A 5-mm specimen of steel *LA* was annealed *in vacuo* at 700° C for 4 h and slowly cooled. After slow bending at 750° C and quenching, ridge veins were present just as in steel *LA*, not vacuum annealed (*see* Fig. 2). Hence, hydrogen does not seem to be a cause of ridge veining.

Since Hanemann and Schrader failed to find the 'veining constituent' in silicon steel[2] the following test with steel *S* (0·05% C, 3·53% Si) was made. After annealing for ½ h at 1000° C and furnace cooling, the structure contained ferrite of two grain sizes (A.S.T.M. 2, elongated; and A.S.T.M. 8) and some pearlite.

A specimen thus treated was bent rapidly at 730° C and quenched. It showed profuse ridge veining in the smaller ferrite grains and some in the large grains too, mostly near their boundaries. Hence silicon does not seem to prevent the formation of ridge veins.

Titanium, when present in sufficient amount, will remove nitrogen and carbon from solution in the ferrite. In an earlier investigation Josefsson and Nelson[8] determined, by means of internal friction measurements, the dissolved carbon and nitrogen contents of steels *P* and *R* upon quenching from 700° C. The dissolved carbon content was found to be about 0·005% for steel *P* and nil for steel *R* and the dissolved nitrogen content to be nil for both steels.

A normalized specimen of steel *P* (grain size A.S.T.M. 6) was bent slowly at 750° C and quenched. The structure of the heavily deformed region showed an irregular network of ridge and groove veining.

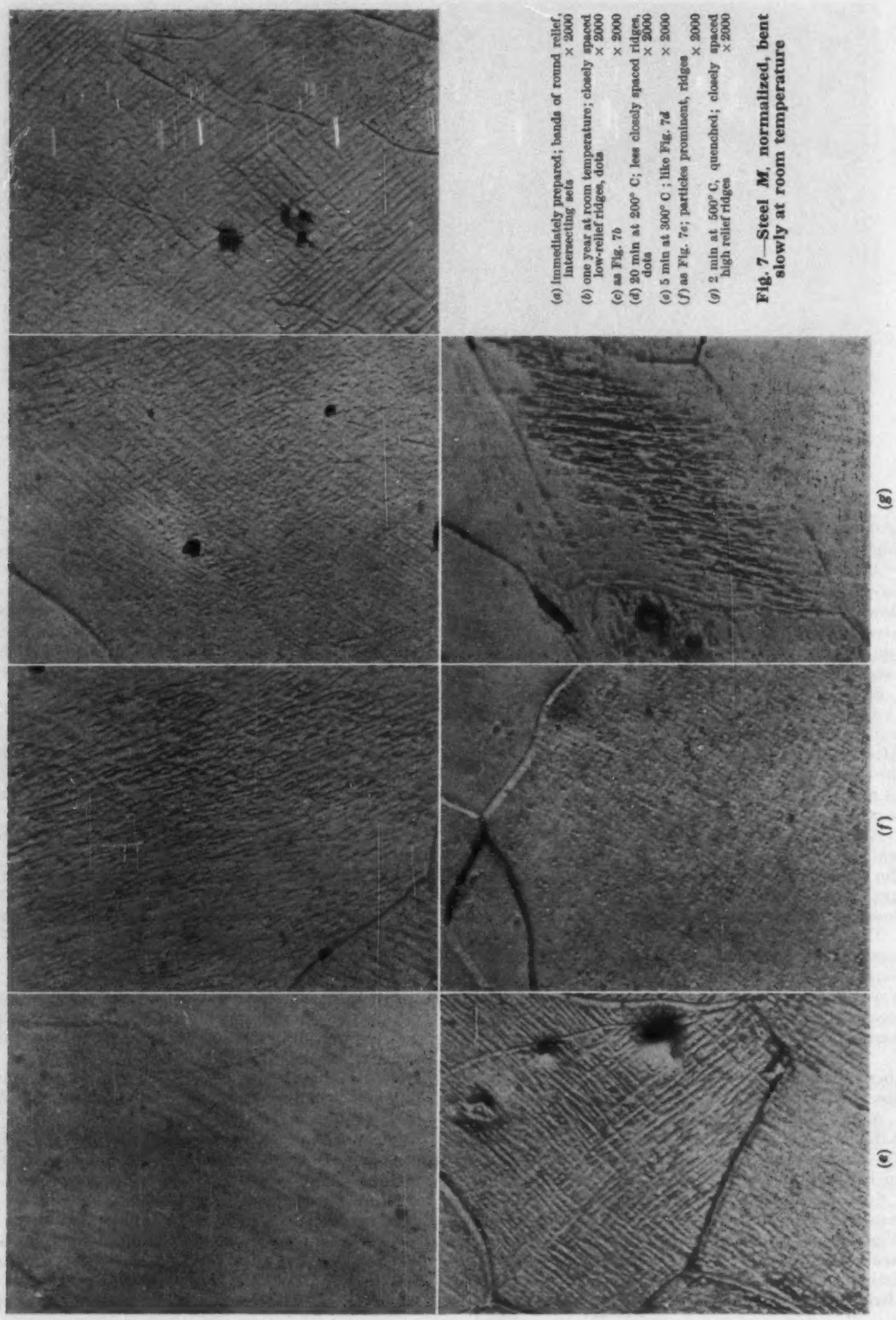
Steel *R* similarly treated (grain size A.S.T.M. 3) showed level veining in bands but no ridge veining (Fig. 8) or ridges at grain boundaries. In both steels fine particles were present, presumably consisting of titanium carbide and nitride or carbonitride. A specimen of steel *R* of finer grain size (A.S.T.M. 6), after bending at 700° C, also showed band level veining only, the band pattern being, however, more irregular as in Fig. 3.

The results thus obtained for titanium steels strengthen the assumption that certain minimum amounts of carbon or nitrogen, or both, in solution may be required for the formation of ridge veining. Accordingly, annealing treatments in various atmospheres were tried on low-carbon steels, with a view to adjusting carbon and nitrogen, separately and in combination, at very low and definite moderate concentrations. The atmospheres used were moist hydrogen for removing carbon and nitrogen, hydrogen saturated with *n*-heptane for introducing carbon, and hydrogen–ammonia mixtures of controlled compositions for introducing nitrogen. Details are given in Table II.

During these treatments, the oxygen content probably also changed but the effect of oxygen alone has not been studied. Since, as will be shown, annealing in moist hydrogen can remove the tendency to ridge veining, it seems unlikely that oxygen could be one cause of this tendency.

Steel *L* was subjected to a series of treatments, details of which are given in Table II with results as to composition given in Table I. As the carbon content decreased in the series *L*, *LA*, and *LC*, from 0·008 to 0·003%, with nitrogen nil, the amount of ridge veining after hot deformation at 750° C decreased too but *LC* still showed some. *LG*, decarburized and recarburized to 0·008% C, showed considerable ridge veining in the heavily deformed regions. *LH*, with 0·003% C and 0·005% N, showed some ridge veining but less than *LG*.

From these experiments it could be concluded that more efficient purifying treatments would probably be required for removing the last remnants of a tendency to ridge vein formation on hot deformation. Accordingly, in the continued studies on steel *K*,



(a) Immediately prepared; bands of round relief, intersecting sets × 2000
(b) one year at room temperature; closely spaced low-relief ridges, dots × 2000
(c) as Fig. 7b × 2000
(d) 20 min at 200° C; less closely spaced ridges, dots × 2000
(e) 5 min at 300° C; like Fig. 7d × 2000
(f) as Fig. 7e; particles prominent, ridges × 2000
(g) 2 min at 500° C, quenched; closely spaced high relief ridges × 2000

Fig. 7—Steel *M*, normalized, bent slowly at room temperature

Fig. 8—Steel *R*, normalized, bent slowly at 750° C, and quenched; band level veins, no ridge veins × 1200

Fig. 9—Steel *KBX*, normalized, bent slowly at 700° C, and quenched; one grain with high-relief ridge veins, one "passive" grain with groove veins and pits × 1500

higher annealing temperatures and various methods for purifying the hydrogen were tried (Table II). The results of bending tests with purified, as well as with subsequently carburized, or nitrided, specimens are shown in Table III. It is seen that, in the absence of nitrogen, ridge veins are first detected when the carbon content is about 0·002% and that as little as 0·005% C causes considerable ridge veining. Unfortunately, the effect of nitrogen on ridge veins is to some extent masked by the effect of 0·002–0·004% of carbon. In the presence of this amount of carbon, 0·003% N has no noticeable effect; with 0·008%, ridge veins are plainly seen; and with 0·025% they appear in amounts comparable to specimen *KAG* with 0·005% of carbon. It has not yet been possible to study the effect of nitrogen in the absence of carbon. Obviously, carbon is the more potent ridge-vein builder of the two. The appearance of ridge veins in steels of dominant nitrogen content was not noticeably different from that of such veins in steels of dominant carbon content.

Table III
BEND-TEST RESULTS

Specimen	C, %	N, %	Deformation Temp., °C	Veining
KAK	0·002	Nil	750 slow	Band level veining, no ridge veins
KAA	0·002	,,	750 fast	Band level veining, traces of ridge veins
KAN	0·003	,,	750 slow	Band level veining, a few short ridge veins
KAG	0·005	,,	750 ,,	Band level veining, plenty of ridge veins
KBS	0·002	,,	750 ,,	Band level veining, no ridge veins
KBU	0·002	,,	700 ,,	Band level veining, no ridge veins
KBZ	0·003	0·003	700 ,,	Band level veining, traces of ridge veins
KCJ	0·004	0·008	700 ,,	Band level veining, combined with short ridge veins
KAJ	0·003	0·025	750 slow	Band level veining, plenty of ridge veins
KBR	0·003	0·025	750 ,,	Band level veining, plenty of ridge veins
KBX	0·003	0·025	700 ,,	Band level veining, plenty of ridge veins (Fig. 9)

Removing and Re-introducing Ridge Veins in Hot-deformed Specimens by Surface Reaction

A specimen of steel *K* (0·020% C, 0·002–0·003% N), normalized from 950° C (grain size A.S.T.M. 7), was bent slowly at 750° C, sectioned, and examined. Most grains in the heavily deformed regions showed a network of ridge veins of different relief together with groove veins (Fig. 10), and also some level veining.

The sectioned specimen was annealed for 25 h at 650° C in hydrogen that had passed over KOH and then through distilled water at room temperature. During the last 5 min the distilled water was by-passed to prevent oxidation of the specimen surface on moving the specimen to the cold part of the furnace. After slight repolish and etching, the structure shown in Fig. 10 had changed to that in Fig. 11 showing level and some groove veining, but ridge veining was entirely absent.

Another specimen of steel *K* that had been decarburized was bent slowly at 750° C and sectioned. It contained level and some groove veining, but no ridge veins. It was annealed for 24 h at 650° C in hydrogen with *n*-heptane. After repolishing and etching the previously prepared section, certain grains showed profuse ridge veining (Fig. 12). After a subsequent decarburizing annealing treatment at 625° for 120 h, the specimen was repolished again and etched: no ridge veins could be seen. These two treatments were then repeated in succession, with the same results.

In the experiments recorded above some grain-boundary migration occurred during annealing but this may be disregarded as not affecting the main theme. These tests show that more or less stationary sub-boundaries can be turned into ridge veins or into level (or possibly groove) veins by carburizing and decarburizing treatments respectively.

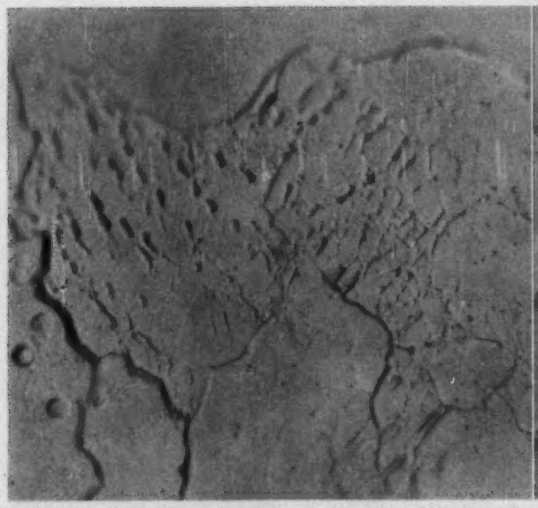


Fig. 10—Steel *K*, normalized, bent slowly at 750° C, and quenched; ridge, groove, and level veins × 1250

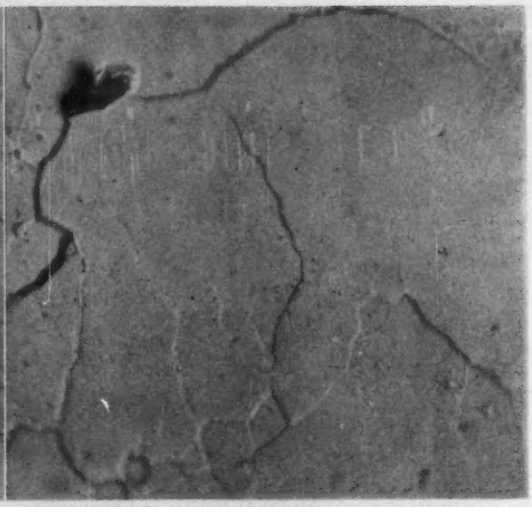


Fig. 11—Same area as Fig. 10, after annealing specimen in moist hydrogen, slightly repolished; level veining, no ridge veins × 1250

INTERNAL FRICTION STUDIES

The results presented in the two previous sections suggest that the formation of ridge veins presupposes the enrichment of carbon or nitrogen, or both, at sub-boundaries. It was hoped that internal friction measurements would provide quantitative evidence of this effect. Accordingly, by methods similar to those previously described, wire specimens with different carbon and nitrogen contents were prepared from steel *T*. Internal friction measurements according to Kê[9] were carried out on specimens from the same lots, upon quenching from 700° C as well as upon stretching 15% at 700° C followed by quenching. The hot-deformed wires showed good correlation between the amount of ridge veins and the concentration of carbon or introgen. However, the hot deformation had no noticeable effect on the internal friction.

Hence it appears that at the sub-boundaries formed during hot deformation only a negligible amount of the interstitial solutes will be concentrated on measuring in the quenched condition. Obviously, sub-boundaries do not have the marked effect of trapping carbon atoms as has been demonstrated to be the case for ferrite grain boundaries.[10] This point requires further study.

INFLUENCE OF ORIENTATION OF SECTION SURFACE ON TYPE OF VEINING RELIEF IN FERRITE

Passivity

As is well known, etching, particularly with nital, will attack the grains in a section of polycrystalline ferrite differently according to their crystal orientation. In the hot-deformed specimens investigated, some grains were always passive, in the sense that the matrix was not, or only slightly, attacked. Veins present in such grains were, however, attacked and always appeared as groove veins. In addition, fine or coarse pitting was generally observed in such grains (Figs. 9 and 13).

In the course of many observations, it became apparent that grains with level veining, in spite of a sometimes pronounced difference in level between sub-grains, were on the whole only moderately attacked and sometimes contained a passive portion (Fig. 14). Grains with ridge veining, with or without groove veining, were, on the other hand, generally heavily attacked. The rate of etching was found to depend upon composition in such a manner that steels of very low carbon and nitrogen contents required longer etching times to give the desired definition of structural features such as veining. In consequence, passive grain sections were more easily recognized by contrast in steels of somewhat higher carbon and nitrogen contents.

In order to determine the orientation of passive grain sections, the following X-ray studies were performed. Specimens were taken from a 2-mm thick plate of electrolytic iron that had been subjected to annealing in moist hydrogen at 1200° C for 4 weeks leading to a coarse-grained structure. The specimens had been slightly carburized in hydrogen with *n*-heptane at 675° C for 7 h. They were not hot deformed.

Figure 15 shows an area consisting of one passive grain, with many veins, to which the incident beam was directed. Test data were: white Co radiation, 40 kV; diaphragm, 0·5 mm; diameter of irradiated area, 1 mm; back reflexion flat camera; film parallel to specimen surface; distance, 3 cm.

Figure 16 shows the Laue pattern obtained. Each diffracted beam had split into three. With the aid of a Greninger net it was calculated that (100) planes of the sub-grains, or rather families of sub-grains producing the spots of the pattern, deviated 8°, 7·5°, and 4° from the prepared surface. In another specimen a passive surface, also with many sub-boundaries, showed deviations of less than 2° from (100).

Hence, it may be concluded that passive grain

Fig. 12—Steel *K*, decarburized, bent slowly at 750° C, and quenched, sectioned, (groove and level veins), then annealed in hydrogen + n-heptane for 24 h at 650° C, same section repolished; profuse ridge veining × 800

Fig. 13—Steel *KBR*, bent slowly at 750° C, and quenched; passive grain section, groove veins and fine pits × 1200

Fig. 14—Steel *KBS*, bent slowly at 750° C, and quenched; grain section in part passive and in part showing banded level veining × 1200

sections coincide more or less with a (100) plane. Engell,[11] who studied electrolytic dissolution of single crystals of α-iron in dilute acids, found the same kind of anisotropy at high current densities. At low current densities, however, Engell found no anisotropy in rate of dissolution.

Similar tests on hot-deformed ferrite, showing level, ridge, and groove veining, gave no spot patterns, only background blackening. Probably the many and considerably smaller sub-grains present after hot deformation have closely scattered orientations, although, as in band level veining, they give the impression of having grouped into a few orientations only. This point requires further study.

For studying the effect of section orientation on the type of veining relief brought out by etching, three specimens of steel *KC* (0·028% C, nitrogen-free) were hot deformed and then sectioned so as to expose two or three sections of each grain.

Specimen *KCH* was slowly bent at 720° C and quenched. Sections I and II, perpendicular to each other, were prepared as shown in Fig. 22*a*. After electrolytic polish and etching, five grains were examined in both sections. The types of veining observed were:

Grain	Section I	Section II
1	Passive surface, groove veins, pits	Ridge veins, different relief
2	Ridge veins, different relief	Passive surface, groove veins, a few pits
3	Ridge veins, different relief	Passive surface, groove veins, a few pits
4	Ridge veins shallow groove veins	Ridge veins, different relief
5	Ridge veins, one set	Ridge veins, different relief

The two sections of grain 1 are shown in Figs. 23 and 24.

Specimen *KCL* was slowly bent at 700° C and quenched. In addition to sections I and II as described for *KCH*, a third section III was prepared at the bend forming an angle of 45° with the other two (Fig. 22*b*). Several grains along the edge II/III were examined, with the following results:

Grain	Section II	Section III
1	Ridge and groove veins	Band level veins
2	Band level veins	Ridge and shallow groove veins
3	Level veins, some groove veins	Ridge and groove veins
4	Groove veins, one set	Groove veins, one set
5	Groove veins, two sets	Band level veins
6*	Passive surface, groove veins, a few pits	
6†	Band level veins	Ridge and groove veins

*away from edge. † at edge

The 45° angle sectioning of this specimen usually permitted the identification of a vein in section II with another vein in section III meeting the former at the edge. Level veins of one section coincided with groove veins of the other in grains 1, 5, and 6; level veins of one section with ridge veins of the other in grain 2; and groove veins in both sections in grain 4. Schematic representations of these relationships are given in Fig. 25. In grain 6, section II was somewhat rounded from the electrolytic polish, hence groove veins of a passive portion gradually passed into band level veins near the edge, an indication that the band level surface deviated little from (100).

Specimen *KCN* was bent slowly at 700° C and quenched. Section I was prepared as before, and sections IV and V perpendicular to it (Fig. 22*c*). Where the sections met in a point they all showed ridge veins of high and low relief.

These studies, although incomplete, indicate that, in the presence of a sufficient amount of dissolved carbon to enable ridge veins to form on hot deforma-

Fig. 15—Electrolytic iron, coarse grained, X-ray target; passive grain section with many groove veins × 150

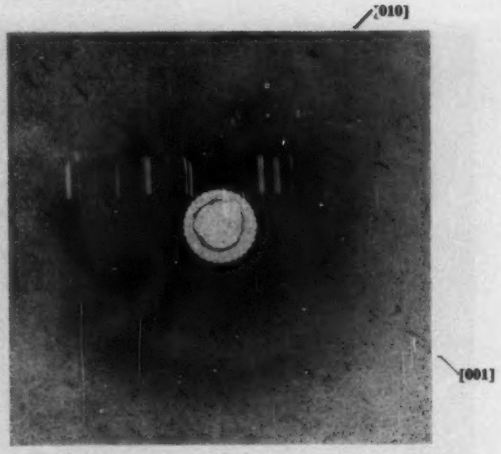


Fig. 16—Laue pattern of grain section Fig. 15

tion, the same sub-boundaries may, in different sections, appear as:

(i) Level and groove veins
(ii) Level and ridge veins
(iii) Groove and groove veins
(iv) Ridge and ridge veins.

Groove veins in a passive section may coincide with any kind of veins in another. In grain sections that were not passive, coincidence between ridge veins in one section and groove veins in another has not been observed. This possibility should not, however, be ruled out.

STRUCTURAL CHANGES DURING AGEING AND ON HEATING

Ageing and heating experiments were carried out on non-deformed, hot-deformed, and cold-deformed specimens in order to find out how different types of veining responded to such treatments, how veining and precipitation reactions were related, and to what extent veining could be produced on heating cold-worked ferrite.

Non-deformed Condition

A series of specimens of steel *M* (0·017% C, 0·002% N), normalized and exhibiting a moderate amount of veining, were heated to 715° C, held for 10 min. and quenched. After ageing at room temperature for one year (Fig. 26*a*), ridge and groove veins remained and a fine precipitate had formed, uniformly distributed and preferentially located at the ridge veins, where precipitate particles could be seen in relief on top of the veins. Ageing at 100° C for 1 h had a similar effect. On ageing at 200° C for 1 h (Fig. 26*b*), the ridge veins remained and the precipitate had coarsened.

The nature of the precipitate formed in ferrite, on ageing at different temperatures, varies with the carbon and nitrogen contents.[12] It has not been studied here.

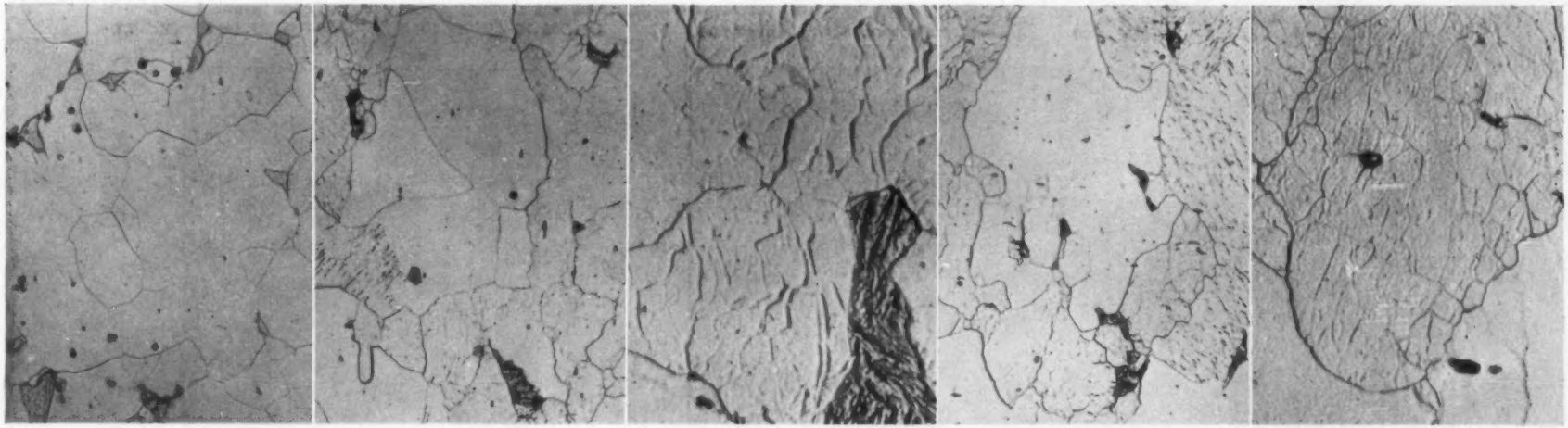
Hot-deformed Condition

One series of specimens of steel *M*, normalized, were bent slowly at 750° C and quenched. On ageing for one year at room temperature, the structure seen in Fig. 18*a* changed to that shown in Fig. 18*b*. The high-relief ridge veins remained but the groove veins had changed into low relief ridges. Dots of precipitate had also formed. After 1 h at 200° C the precipitate was heavier (Fig. 18*c*) but there was no other change.

Another series of specimens of steel *M*, normalized, were bent rapidly at 750° C and quenched. As quenched and after ageing for one year at room temperature, the specimens exhibited structures shown in Figs. 19*a* and 19*b* which, in regard to groove veining and dots are counterparts to Figs. 18*a* and 18*b*. After ageing at 100° C for 15 min the structure was as shown in Fig. 19*d*. Ridge and groove veins remained and a fine precipitate was seen, in places modifying the appearance of the former. During ageing at 100° C for increasing periods, the groove veins gradually disappeared and were replaced by low-relief ridges as seen in Fig. 19*e* for an ageing period of 70 h.

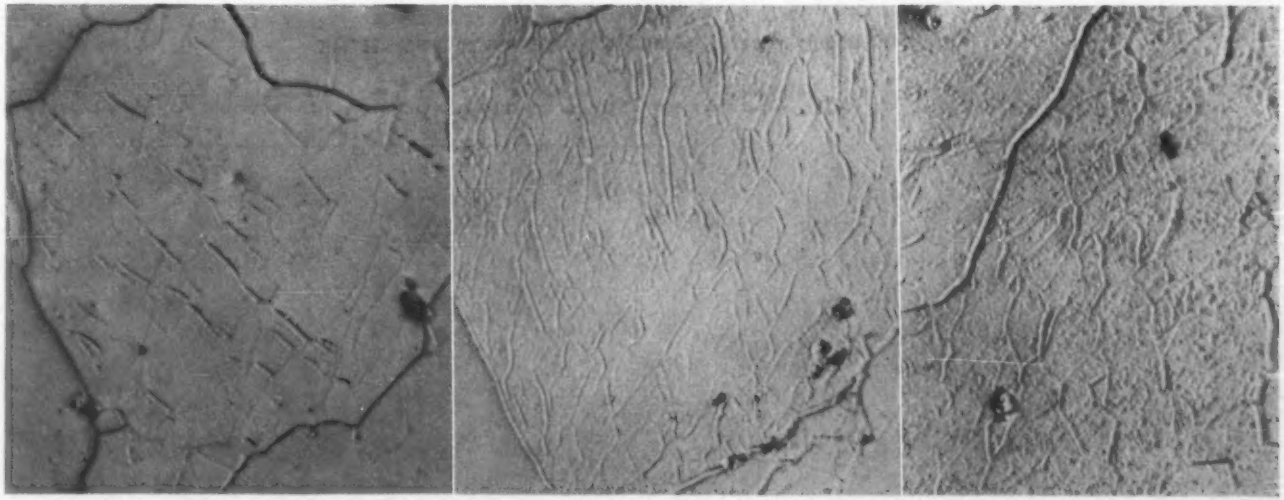
Specimens of steel *K* (0·020% C, 0·002–3% N), normalized, were bent rapidly at 700° C. After this treatment ridge and groove veins were present. On heating one specimen at 200° C for 4 h, the only veins present were ridge veins and low ridges (Fig. 21*a*). A series of specimens that were similarly aged at 200° C were reheated at 300°, 400°, 500°, 600°, and 700° C and quenched, the holding times being 60, 30, 15, 10, and 5 min respectively. After treatment at 300° and 400° C no groove veins were observed. After treatment at 500° C traces of groove veins could be seen, whereas the structures obtained after treatment at 600° and 700° C exhibited groove veins (Fig. 21*b*) to the same extent as before ageing at 200° C. In the whole series of specimens the amount of high-relief ridge veins appeared to be unchanged.

Apparently, there is an essential difference between ridge and groove veins. The former, once they have formed, cannot be removed by heating alone, barring recrystallization and possibly prolonged treatment



(*a* middle region; smooth grain boundaries, little veining × 300) (*b*) stretched region; ragged grain boundaries (except for recrystallized grains), profuse veining × 300) (*c*) stretched region; ridge veins of high and low relief × 1200) (*d*) compressed region; similar to Fig. 17*b* × 300) (*e*) compressed region; ragged boundaries of grains with ridge veins of high and low relief, grain with level veining, recrystallized grain × 1200

Fig. 17—Steel *M*, normalized, bent rapidly at 850° C, and quenched



(*a*) immediately prepared; ridge and groove veins, openly spaced × 2000) (*b*) one year at room temperature; high-relief ridge veins and low-relief ridges, but no groove veins, dots × 2000) (*c*) 1 h at 200° C; similar to Fig. 18*b*, but dots more marked × 2000

Fig. 18—Steel *M*, normalized, bent slowly at 750° C, and quenched

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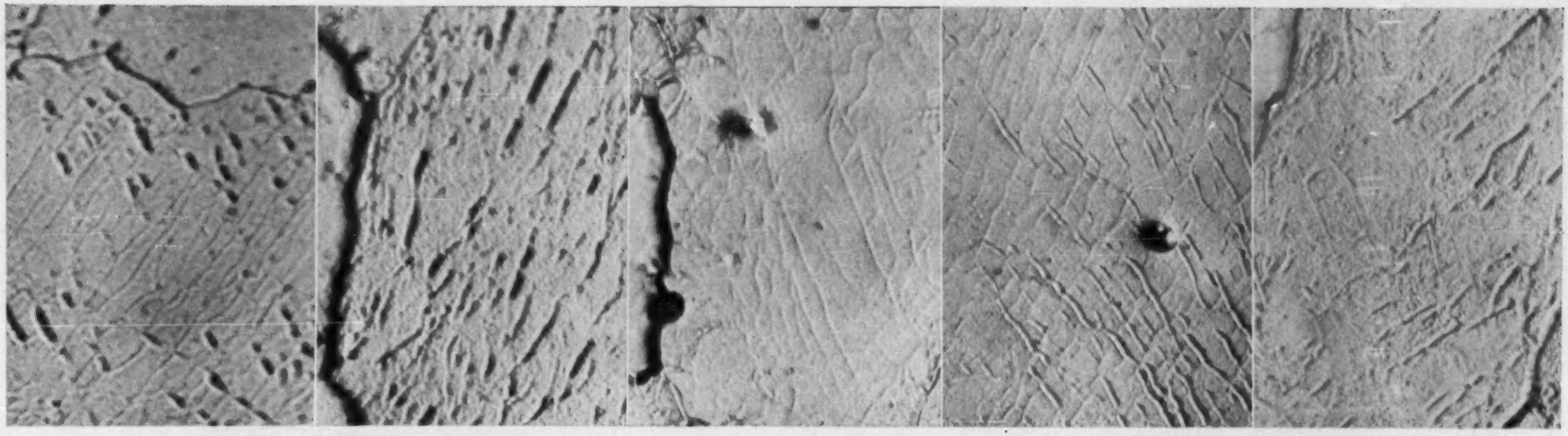


(*a*) immediately prepared; ridge and groove veins, closely spaced, two intersecting sets × 2000

(*b*) one year at room temperature; ridge veins but no groove veins, dots × 2000

(*c*) 15 min at 100° C; groove veins, intersection sets × 2000

(*d*) 15 min at 100° C; ridge and groove veins, dots × 2000

(*e*) 70 h at 100° C; high-relief ridge veins, low-relief ridges, dots × 2000

Fig. 19—Steel *M*, normalized, bent rapidly at 730° C, and quenched

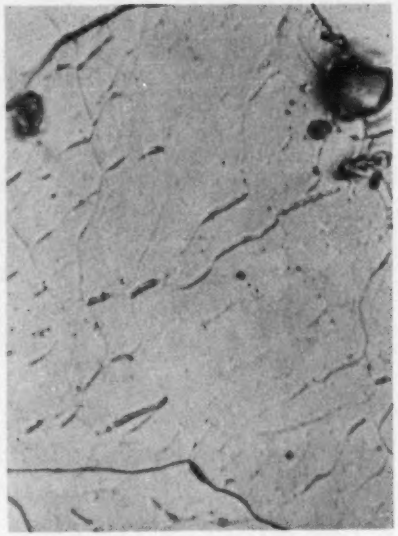
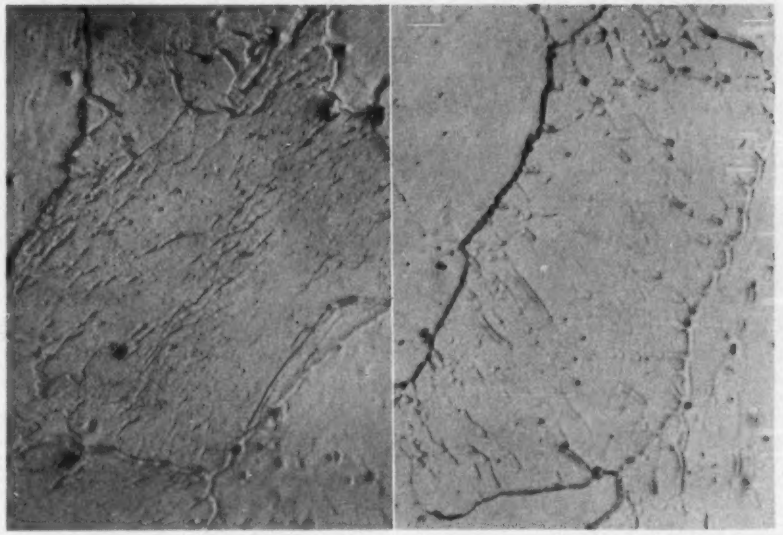


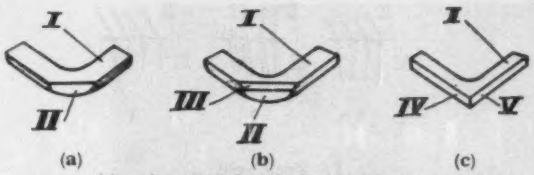
Fig. 20—Steel *M*, normalized, bent slowly at 700° C, and furnace cooled; high relief ridge veins and low relief ridges × 2000



(*a*) 4 h at 200° C; ridge veins and low ridges × 1500

(*b*) as Fig. 21*b*, then 10 min at 600° C; ridge veins and again groove veins × 1500

Fig. 21—Steel *K*, normalized, bent rapidly at 700° C, and quenched



(*a*) specimen *KCH*, bent slowly at 720° C and quenched
(*b*) „ *KCL*, „ „ „ 700° C „ „
(*c*) „ *KCN*, „ „ „ „ „ „

Fig. 22—Sketch showing sectioning of specimens

that causes sub-boundaries to move. Groove veins, on sufficient ageing, undergo transition to low-relief ridges, which are, however, restored to groove veins on heating to higher temperatures and quenching. The formation of low-relief ridges on ageing probably requires a certain amount of dissolved carbon or nitrogen or both.

The existence of groove veins in certain hot-deformed steels after quenching (or possibly air cooling) but not slow cooling, when low-relief ridges are found, probably implies that an ageing effect, such as described, may occur during slow cooling.

Cold-deformed Condition

Specimens of steel *M* (0·017% C, 0·002% N), normalized, were slowly bent at room temperature. Figure 7*a* shows the structure of a heavily deformed ferrite grain. Etching has brought out two sets of intersecting bands of rounded relief. Figures 7*b* and 7*c* show the effect of ageing at room temperature for one year. There are closely spaced, interrupted, and continuous low ridges and also round elongated particles that may or may not differ in character from the ridges. A similar structure was found after ageing at 100° C for 1 h and at 200° C for 20 min (Fig. 7*d*). After 5 min at 300° C (Figs. 7*e* and 7*f*), precipitate particles are prominent and intersecting continuous ridges are still present. The structure was similar after 2 min at 500° C. In certain grains (Fig. 7*g*) the closely spaced ridges were similar to those found after hot deformation at 550° C (Fig. 6). After 2 min at 600° C, closely spaced ridges with high relief were also found.

Hence, the effect of ageing cold-deformed steel *M* appears to be replacement of the band structure of Fig. 7*a* by sets of more closely spaced narrow ridges with low and varying relief and precipitation of particles.

Specimens of the low-carbon steel *LA*, normalized, were slowly bent at room temperature and then heated. After holding at 600° C for 2 min, one specimen displayed intersecting, fairly wide bands of alternating level, with a few short ridges along band boundaries. This structure resembled that of rapidly hot-deformed low-carbon steel (cf. Fig. 1).

The ridges of high relief, observed after heating cold-deformed steels *M* and *LA* at 500° C and higher, are believed to be of the same kind as the so-called ridge veins in hot-deformed steel.

DISCUSSION

Our results confirm earlier findings that veining is caused by hot deformation.[1] This effect has been disputed by Samuel and Quarrell[6] whose concept of

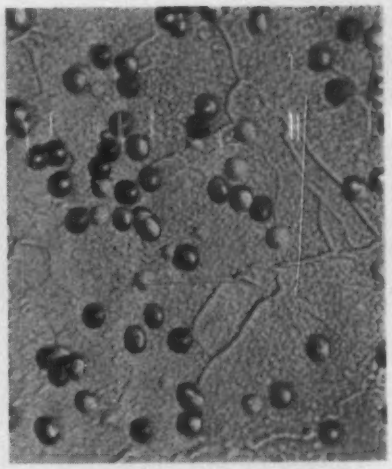


Fig. 23—Specimen *KCH*, section I (Fig. 22*a*); passive grain section, groove veins, pits × 1000

a-veining does not seem to cover sub-boundaries in general, but only sub-boundaries clearly marked by ridges or precipitates. They claim that *a*-veining is formed during sub-grain formation caused by strains set up by the $\gamma \rightarrow a$ transformation. Hultgren and Herrlander[1] regard this as a special case of hot deformation and, as pointed out in the discussion[13] on the paper by Samuel and Quarrell, there is no reason to distinguish between veins resulting from hot deformation caused by different means.

Formation of Sub-boundaries by Hot Deformation

The mechanism by which veins or sub-boundaries are formed in ferrite during hot deformation is probably complicated. It may be assumed that polygonization plays an important role in this process, in a similar manner to that observed in creep tests.[5]

On increasing the rate or lowering the temperature of deformation, the veining pattern becomes more closely meshed and the tendency for parallel sub-boundaries to form increases. Increasing the grain size has a similar effect. Level veining in such cases often displays bands of alternating level, presumably reflecting alternating orientations. Possibly this results from some sort of kinking similar to the deformation bands found by Barrett and Levenson[14] in cold-deforming iron.

Orientation Effects and Etching Properties of Sub-boundaries

The sub-boundary topography resulting from etching will depend on the orientation of the grain section. This can partly be explained in terms of the anisotropic etching characteristics of ferrite. When the grain section is close to (100) the rate of etching of the matrix is slow (passivity). The lattice imperfections of existing sub-boundaries will then promote the rate of attack and, in consequence, grooves will be formed. In grain sections sufficiently deviating from (100), etching will be more or less rapid, when some sub-boundaries may appear as grooves and others as ridges. The arguments used for grooves in passive sections are believed to apply here also.

Fig. 24—Specimen *KCH*, section II; ridge veins, different relief × 1000

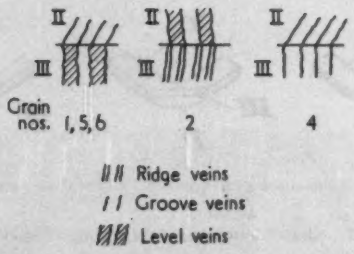


Fig. 25—Specimen *KCL*, schematic representations of veining in sections II and III of the same grain

The ridges show a higher resistance to etching than does the matrix. This effect has been found to be due to interstitial solutes, carbon and nitrogen, and will be discussed later. Level veining implies that the sub-boundaries are not preferentially attacked by the etchant and they become noticeable only because of variations in relief, owing to slightly differing sub-grain orientations. There is evidence that grain sections exhibiting level veining have orientations not far from (100). Level veining may also occur together with ridge and groove veins. Such grain sections are believed to have an orientation intermediate between those of the two types described.

It has been pointed out that, in grain sections showing level veining, differences in level between adjacent sub-grains are often prominent, in spite of only moderate general depth of etching. On the other hand, in grain sections with ridge and groove veins only, which are always deeply etched, such differences in level have not been observed. It may be that a given increment of change in orientation will influence the rate of etching far more strongly when the section is close to (100).

Apart from the effect of anisotropic etching of the matrix, it may be expected that the etching of sub-boundaries will also depend on the orientation of the sub-boundary with respect to the specimen surface. Owing to the finite number of operative slip systems, a certain coupling between these two effects is likely.

In an etched section, ridge and groove veins have an apparent minimum width of the order of 0·2μ. The thickness of a sub-boundary, as defined by the range affected by its lattice imperfections, is considerably smaller.[15] It would appear that some kind of exaggeration is involved in the etching process.

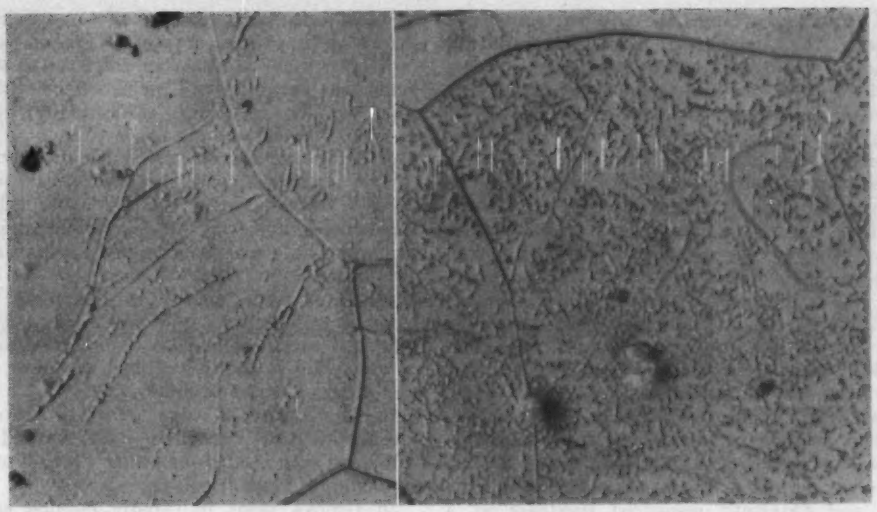
Nature of Ridge Veins

Apparently the phenomenon appearing as ridges upon etching is rapidly developed at the deformation temperature or during the subsequent cooling. Thus it has not been possible to suppress the formation of ridges by rapid quenching, nor has holding at temperature after deformation affected the amount of ridge veins present. In the authors' opinion the only way to remove ridge veins without changing the composition of the ferrite is by recrystallization in the α-range which implies the destruction of all sub-boundaries. Samuel and Quarrell[6] state that veining networks can also be eliminated by quenching from temperatures in the range 700–850° C, but will persist upon slow cooling from this range. Judging from numerous observations made during the present investigation, it seems probable that in many cases Samuel and Quarrell regarded precipitates located at sub-boundaries as '*α*-veining', while veins present in quenched specimens (and hence free from precipitates) may have been overlooked.

That part of the veining, formed upon hot deformation, which appears as ridge veins is clearly related to the carbon and nitrogen contents of the ferrite.* When carbon is present alone in a concentration of less than 0·002%, ridge veins have not been observed. However, at 0·005% C they are already quite prevalent and they occur to a still larger extent at higher carbon contents. The fact that ridge veins in a grain section can be removed and reintroduced, by means of decarburization and carburization respectively, also illustrates the effect of carbon in giving rise to ridge veins. Nitrogen has a similar influence although less pronounced. This statement is true with the reservation that carbon in concentrations in the range 0·001–0·004% has been present in the nitrided specimens investigated. The observations on the titanium-stabilized steel *R* provide conclusive evidence that carbon or nitrogen in solution are needed for ridge vein formation.

As a consequence of these findings it is assumed distinguished from ridge veins formed on hot deformation. After heating to a temperature above the range

* In a recent paper Wever, Fischer, and Engelbrecht[16] studied the effect of oxygen on the microstructure of iron. Specimens from ingots, of varying oxygen contents, that had solidified and slowly cooled in the moulds all showed considerable veining. It was suggested that the veining was a result of the $\gamma \rightarrow \alpha$ transformation and had no connection with the oxygen content. By private communication with Dr. Wever and Dr. A. Schrader it was learned that the carbon and nitrogen contents in the various melts were sufficiently high, judging from the results of the present investigation, to account for the ridge veins.



(*a*) one year at room temperature; ridge and groove veins, precipitate particles uniformly distributed and also preferentially located at ridge veins × 2000
(*b*) 1 h at 200° C; similar to Fig. 26*a* but coarser particles × 2000

Fig. 26—Steel *M*, normalized, heated at 715° C for 10 min, and quenched

that ridge veins are due to an enrichment of carbon, or nitrogen, or both, at certain sub-boundaries. Since some portions of the sub-boundaries within a grain section do not appear as ridges, it is obvious that carbon and nitrogen in solution do not by themselves necessarily give rise to ridge-vein formation. Observations indicate that in large grains ridge veins are more prevalent in areas adjacent to grain boundaries, where complex deformation processes are likely to occur and hence the misfit at sub-boundaries would be more pronounced. Furthermore, grain boundaries, where in general the degree of misfit is also high, very often display ridges. It seems plausible that the latter are similar in nature to ridge veins. However, as pointed out, internal friction measurements indicate that the enrichment of carbon or nitrogen at sub-boundaries is very small, whereas the enrichment at grain boundaries is considerable in the case of carbon and very small in the case of nitrogen.[10] During the present investigation it has also been observed that, on repeated etching in nital, grain-boundary ridges often display a terraced appearance, whereas this has not been observed for ridge veins. The significance of these dissimilarities between sub-boundaries and grain boundaries is not clear.

The enrichment of interstitial atoms at sub-boundaries may be the result of a Cottrell mechanism involving the formation of atmospheres of solute atoms round the sub-boundary dislocations. If this were so, one would expect the amount of ridge veins, as related to the degree of enrichment, to depend on temperature in the same way as is proposed for the density of atmospheres.[15] By efficient quenching from a temperature sufficiently above the temperature of condensation (estimated[15] to 700° K) it should be possible to eliminate ridge veins, and by ageing at moderately low temperatures the amount of ridge veins should increase. The fact that such effects have not been observed could then only be explained by the rapid and complete development of atmospheres during the course of cooling. There remains the question why only portions of the sub-boundaries provide locations for atmospheres. This cannot be due to a shortage of solutes since, according to the internal friction measurements, the quantity of interstitial atoms trapped when forming profuse ridge veining seems to be very small.

Tsou, Nutting, and Menter[17] using electron diffraction obtained evidence for austenitic grain-boundary ridges in an ingot iron containing 0·026% C and 0·006% N. According to the lattice parameter, the carbon or nitrogen content of the austenite was estimated to be about 1%. This phenomenon has been observed in later investigations[18, 19] and offers an alternative explanation of the nature of ridge veins. The latter could represent austenite films which form on hot deformation at favoured portions of the sub-boundaries.

It does not seem likely that ridge veins are of the same nature as regular carbide or nitride precipitates, since, upon ageing, particles of the latter may be found enclosed within the ridges (Fig. 26*a* and *b*). The nature of ridge veins remains to be established by further studies.

Ageing Effects in Hot-deformed Ferrite

The sub-boundaries that together with ridge veins complete the sub-boundary network in deeply etched grain sections may, upon quenching, appear as groove veins. Much care has to be taken in bringing out such veins by etching. Groove veins have been observed at carbon and nitrogen concentrations sufficiently low for ridge veins to be absent. If the specimen is slowly cooled or quench-aged on hot deformation the groove veins are obliterated or transformed to ridges of low relief. The latter are easily

500–600° C and quenching, the groove veins will again appear. This behaviour suggests that during furnace-cooling or quench-ageing the groove veins will provide sites for precipitates, or possibly Cottrell atmospheres, which will dissolve on reheating to sufficiently high temperatures contrary to high-relief ridge veins. Alternatively, the slowly cooled ferrite may contain a fine precipitate that would cause the rate of etching of the matrix to be comparable to, or faster than, that of the sub-boundaries, whereas the absence of such a fine precipitate—not formed in quenching or dissolved in reheating—in specimens quenched above 500–600° C would make the matrix more resistant to etching than the sub-boundaries. As mentioned above, precipitates formed on ageing and visible by microscope are preferentially located in ridge veins. Such precipitates, however, form discrete particles. What is said here regarding groove veins in deeply etched grains does not apply to groove veins in passive sections. The latter will not change appearance during low-temperature heat treatments.

Samuel and Quarrell[6] measured hardness as a function of time during tempering in the range 400–600° C on specimens quenched from 800° C. During these treatments α-veining networks were said to develop and the activation energy for the process, as derived from the hardness curves, was 18 kcal/mole. Since the activation energy for the diffusion of carbon or nitrogen is of the same order of magnitude, Samuel and Quarrell considered their findings as evidence for carbon or nitrogen being the solutes causing α-veining. However, as they started from a supersaturated solution with respect to these elements, their observations regarding the development of α-veining may have been due to the formation of precipitates at sub-boundaries. Furthermore, the activation energy for a macroscopic process like the observed change in hardness can hardly be regarded as simply related to the activation energy for diffusion of interstitial solutes.[20]

Ageing Effects in Cold-deformed Ferrite

Some structures observed after heating specimens deformed at room temperature probably illustrate several interdependent processes. Ridges appearing after ageing at low temperatures (Figs. 7*b–e*) differ from ridges found in hot-deformed and quenched specimens and may result from combined effects of rearrangement of dislocations, atmosphere formation, and precipitation. However, ridges found in cold-deformed specimens after heating at 500° C and higher temperatures (Fig. 7*g*) resemble those found after hot deformation at similar temperatures and may have the same nature.

CONCLUSIONS

(1) Sub-boundaries in ferrite are formed during hot deformation, probably mainly by polygonization. The sub-boundary network tends to show greater regularity of directions and to be more closely meshed the higher the deformation rate and the lower the temperature. A coarse grain size will lead to more regular sub-boundaries than a fine grain size.

(2) Depending on the orientation of the grain section such sub-boundaries appear differently on etching in nital, viz. (*a*) in (100) sections exhibiting the lowest etching rate, always as groove veins; (*b*) in rapidly etching sections as ridge veins and groove veins (the latter in the quenched condition); (*c*) in some sections, apparently close to (100), as difference in level between sub-grains: level veining.

(3) Ridge veins are caused by the enrichment of carbon or nitrogen, or both, at certain sub-boundaries, possibly those having a high degree of misfit. During this process, however, only negligible amounts of the available content of the interstitial solutes are trapped by the sub-boundaries.

(4) Ridge veins, once formed, are not obliterated by heat treatments within the α-range, except by a recrystallization anneal removing the sub-boundaries or annealing in moist hydrogen to lower the carbon and nitrogen contents sufficiently.

(5) The nature of ridge veins is unknown. They are not identical with precipitates formed on ageing. They are suggested to be either Cottrell atmospheres or austenite films at sub-boundaries.

(6) Groove veins in rapidly etching sections are obliterated or transformed to low-relief ridges during ageing or slow cooling. This appears to be a precipitation effect. Groove veins will again appear on quenching from temperatures above the range 500–600° C.

(7) Ageing at low temperatures of cold-deformed ferrite results in the formation of irregular ridges of low relief along closely spaced lines conforming to the deformation pattern. At intermediate temperatures the quantity of such ridges decreases whereas precipitated particles independent of the deformation pattern become more prevalent. These structures are probably caused by complex processes involving Cottrell atmosphere formation, precipitation, and rearrangement of dislocations.

(8) Upon ageing cold-deformed ferrite at temperatures above about 500° C, closely spaced ridges are observed in some deeply etched grain sections. These ridges are believed to form at sub-boundaries and to be of the same nature as ridge veins formed during hot deformation.

Acknowledgments

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A Vacuum Fusion Apparatus for Gas Analysis

By P. D. Blake, M.A.

SYNOPSIS

An apparatus is described which is suitable for the determination of the total oxygen and fractional oxygen content of steels. It employs a compact type of split-graphite resistor. The gas analysis section is integral with the heating unit and comprises a dual system suitable for either low-pressure gas analysis where small volumes of gas are involved, or medium-pressure gas analysis for volumes of gas up to about 12 ml. 1449

Introduction

THE GENERAL PRINCIPLES of vacuum fusion analyses have been well discussed by Sloman[3, 4] and Stevenson and Speight[3]. The objective in modifying the original Speight and Cook design[1] has been to produce an apparatus with greater flexibility and hence more suitable for basic research into the effects of gases upon the properties of steel.

DESCRIPTION OF APPARATUS

The general layout of the present equipment is shown in Fig. 1. The apparatus comprises a furnace section, a calibrated bulb and Toepler pump section, and low- and medium-pressure gas analysis sections together with the appropriate backing lines.

The chief modifications which have been made to the original Speight and Cook apparatus will be described. The split-graphite resistor assembly is very similar to the original except that an additional graphite radiation shield, which is suspended from the top of the funnel, has been added. In place of the stainless-steel tube and quartz head, a single quartz furnace tube has been substituted, thus eliminating a joint in an inconvenient position. The flange on the lower end of the furnace tube is sealed to the furnace baseplate by a silicone-rubber O-ring. Silicone-rubber washers are also used to seal the electrodes to the furnace baseplate. It has been possible to obtain an improved design of diffusion pump with a pumping speed at the flange of 70 l./s and capable of operating against a backing pressure of 35 mm Hg. This means that volumes of gas up to 20 ml can be collected in the calibrated bulb section.

The principal difference between the present apparatus and that of Speight and Cook is the incorporation of a dual analytical system which has certain advantages as follows:

(i) The low-pressure analytical system, in which the method of gas analysis is that of Ransley,[3] is extremely useful for samples of low gas content or of low weight. This section can be used for volumes of gas up to 2 ml. A complete analysis of the gas is possible, including the residual hydrogen content and the methane content

(ii) For gas volumes greater than 2 ml, the medium-pressure gas analysis system is used. In operating this section, a known volume of gas is drawn from the analytical section via tap *G* over the Arneil catalyst[5] at 300° C, through magnesium perchlorate and then through the cold trap *T* at 20° C by diffusion pump *P*III, being finally pumped back into the calibrated bulb section. The volume is measured and the gas re-circulated to constant volume, the decrease in volume being that of the hydrogen. Liquid nitrogen in a thermos flask is then placed on the cold trap *T* and the gas recirculated again to constant volume, this difference in volume being that of the carbon monoxide.

If methane is thought to be present, this can be oxidized after the carbon monoxide by increasing the temperature of the Arneil catalyst to 550° C.

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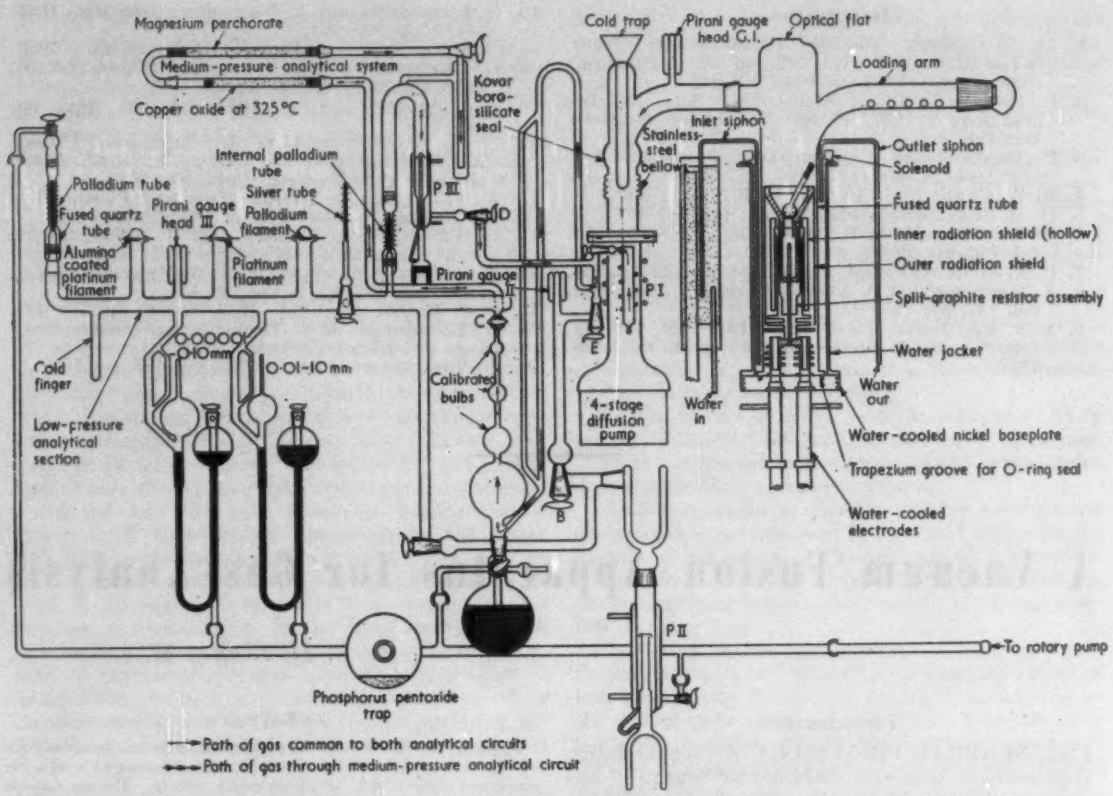

Fig. 1—Vacuum fusion apparatus

Carbon-monoxide/carbon-dioxide pressures up to about 2–3 mm can be handled in this section, 3 mm being the limiting pressure of diffusion pump *P*III. This means that total gas pressures of about 5 mm can be handled, provided the excess over 3 mm is hydrogen, which is converted to water vapour and absorbed by the perchlorate. The time for an analysis is about 15 min.

The blank in the analytical section over a 24-h period is negligible. The average total blank is 0·095 ml per half-hour at 1600° C and is as follows:

0·06 ml carbon monoxide, 0·03 ml hydrogen, 0·005 ml nitrogen.

This compares with a blank of 0·01 ml per half-hour obtained by Speight and Gill[6] and 0·002 ml per half-hour by Sloman,[3] using high-frequency heating.

Table I gives some published blank figures indicating the lower order of blank obtained with high-frequency furnaces; it also shows the higher degassing temperatures available with the latter type of furnace.

This apparatus is also being used for fractional vacuum fusion analysis as described by Swinden, Stevenson, and Speight.[7]

EXPERIMENTAL RESULTS

Table II contains results by the low- and medium-pressure analytical systems respectively on (*a*) billets of rimming steel, and (*b*) weld metal as laid down by five of the classes of electrodes in B.S. 1719 (1951), i.e. classes mainly for welding mild and medium-carbon steels. The all-weld-metal test pieces were prepared in accordance with Appendix C of B.S. 639 using 6-gauge electrodes throughout. The samples were machined from the parallel portion of tensile specimens made in accordance with B.S. 18 or from Izod specimens made in accordance with B.S. 131, a weight of sample between about 5 and 15 g being used according to the gas content.

Discussion of Results

The reproducibility of the results obtained with this apparatus and their absolute accuracy will be dealt with separately, as follows:

Reproducibility—A maximum difference of ± 0·0005% oxygen was obtained using the low-pressure analytical system, as shown by Table II.

The results obtained when using the medium-pressure system are equal in reproducibility to the above on the deposits laid down by B.S.1719 Classes 1, 2, 3, 5, and 6 electrodes, but on the Class 4 weld the results show a larger variation. This may possibly be due to inherent variations in this type of iron-oxide–silicate-coated electrode deposit.

Table I

SOME PUBLISHED BLANK FIGURES FOR VACUUM FUSION APPARATUS

	Cook and Speight[1]	This Apparatus	Sloman[3](b) 1550° C	Speight and Gill[3] 1600–1650° C	Stevenson and Speight[2](a) 1650–1700° C
Total blank, ml/h	0·2	0·2	0·004	0·02	0·4
% by volume: Carbon monoxide	70	63	35	45	7
% by volume: Oxygen	not	trace	trace		6
% by volume: Hydrogen	determined	31	20	45	51
% by volume: Nitrogen	30	6	45	10	36
Type of furnace	Split-graphite resistor			High frequency	
Max. degassing temperature, ° C	2100	2200	2600	2400	2100

Absolute Accuracy—As a criterion of absolute accuracy, it was thought that a comparison with other laboratories would give a reasonable indication of the error involved.

Table III shows that figures obtained with this apparatus using either the low- or medium-pressure analytical systems are within 0·001% oxygen of the figures obtained by the N.P.L. apparatus[8, 10] using the vacuum fusion method. The figures obtained with the medium pressure analytical system are within 0·01% oxygen of those obtained by the alcoholic iodine method by Garside, Rooney, and Belli[9] on welds containing less than 0·1% oxygen, but are 0·025% lower on the high-oxygen weld; the agreement here, however, would not be expected to be as good as between two vacuum fusion equipments. It must also be pointed out that these welds were laid down under different conditions from different batches of electrodes.

SUMMARY AND CONCLUSIONS

A vacuum fusion apparatus is described which, with an output of five samples per run, compares satisfactorily with other types of vacuum fusion apparatus in which the gas analysis unit is a separate unit. The present apparatus precludes any possi-

Table II

REPRODUCIBILITY OF RESULTS IN THE TWO ANALYTICAL SYSTEMS

Steel Analysed	Total Gas, ml	Oxygen, wt-% Result 1	Oxygen, wt-% Result 2	Difference Values	Analysis System
Rimming steel billet, position 1	1·8	0·0130	0·0140	± 0·0005	Low pressure
Rimming steel billet, position 2	1·8–1·9	0·0132	0·0132	nil	Low pressure
Rimming steel billet, position 3	1·7	0·0121	0·0122	± 0·00005	Low pressure
Rimming steel billet, position 4	1·8	0·0132	0·0142	± 0·0005	Low pressure
B.S.: 1719 Class 1 weld	2·85, 3·0	0·064	0·063	± 0·0005	Medium pressure
B.S.: 1719 Class 2 weld	5·7	0·075	0·075	nil	Medium pressure
B.S.: 1719 Class 3 weld	4·8, 5·7	0·090	0·0895	± 0·00025	Medium pressure
B.S.: 1719 Class 4 weld	12·9	0·117	0·121	± 0·002	Medium pressure
B.S.: 1719 Class 5 weld	16·6, 16·8	0·135	0·136	± 0·0005	Medium pressure
B.S.: 1719 Class 6 weld	3·1	0·026(0)	0·026(3)	± 0·00015	Medium pressure

Table III

COMPARISON OF RESULTS WITH OTHER LABORATORIES*

Steel Analysed	% Oxygen (this apparatus)	Analysis System	% Oxygen (alcoholic iodine method)	% Oxygen (Sloman et al. at N.P.L.)
Rimming steel sample †	0·013% 0·013%	Low pressure	...	0·012% (see ref. 10)
B.S.: 1719 Class 2 weld	0·075%, 0·075%	Medium pressure	0·072% (see ref. 9‡)	...
B.S.: 1719 Class 3 weld	0·090%; 0·089%§	,,	0·091%, 0·10% (see ref. 11)	0·09% (see ref. 8§)
B.S.: 1719 Class 5 weld	0·135%	,,	0·16% (see ref. 11)	...

* The electrodes are the same, but not from the same manufactured batch; the welds were laid down under different conditions, but all were U-butt type welds.
† The figure of 0·012% oxygen on the rimming steel sample was obtained by the N.P.L. on a sample submitted to them.
‡ This result refers to arc weld 2 in their paper,[9] details of which were kindly supplied by the authors.
§ This result refers to electrode C in the paper,[8] details of which were kindly supplied by H. A. Sloman.

bility of loss or contamination of the gas during transfer to the analytical unit.

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The Diploma in Technology

The National Council for Technological Awards in its first report states that the Government's plan to produce the necessary supply of technologists through the technical colleges can only succeed if the colleges, industry, and the Council are able to pool their experience and ideas and produce suitable courses.

The report describes the work of the Council since it was set up at the end of 1955 under the Chairmanship of Lord Hives, to create and administer awards of high standard for technical college students.

In his foreword to the Report, Lord Hives states that on 21st November there were 965 students at 11 colleges attending 37 courses recognized as leading to the Diploma in Technology.

As part of the procedure for the approval of courses, members of the Council visited and reported on the colleges concerned in order that a "full and serious examination for a just appraisal" could be made.

"No single college," states the report, "satisfies fully the Council's requirements and in some cases good work is being done in singularly unsuitable accommodation... but the substantial amount of new building either now in progress or planned to start in the immediate future, gives firm promise that the majority of these deficiencies will have been met by the time that the renewal of the recognition of courses is considered."

The Council are anxious to see that all courses leading to the Diploma not only reach the required standard in a technology, but also cover liberal studies and informal activities. To provide for this a large amount of pioneer work has to be done and for the moment the Council have been satisfied to accept a statement of the intention of the college on the development of a liberal approach in the curriculum, rather than to insist on the submission of detailed syllabuses in all cases.

The Council regard a course leading to the Diploma in Technology as a complete entity, the industrial training being just as much a part of the course as the academic study. The report emphasizes that the integration of academic and practical training in industry must be an essential part of any course. The Council also consider it to be of paramount importance that students continue their academic studies during periods of industrial training. Some colleges are attempting to solve these problems by arranging for members of staff to visit students at their work while they are in industry.

Letter to the Editor

Similarity Criteria for Dust-deposition Tests in Models

SINCE THE PUBLICATION of our report in October 1957 (*J. Iron Steel Inst.*, 1957, vol. 187, pp. 121–125) we have received a variety of enquiries from investigators who wish to adopt our suggestion that deposition tests should be carried out in models with a liquid rather than a gas as the fluid medium. This is particularly desirable when modelling a system through which air or other gases are passing at temperatures and pressures near to atmospheric, since if air is used in the model, the tests must be conducted at somewhere near full scale to maintain Reynolds and Froude number similarity simultaneously.

The original calculations related to our work on air models and, since the buoyancy forces on a particle were then less than a thousandth part of its weight, the effects of buoyancy were not included. It will perhaps be of service if we now re-calculate the similarity criteria including buoyancy forces and show that this results in a simple modification to the particle number P and Froude number F.

In addition to the symbols employed in the original report, let the following be used:

ρ_f = Density of the fluid

ρ_p = Density of the particle

m = Mass of the particle *in vacuo*

Thus, re-writing equation (1), the equations of motion are:

$$m\frac{\mathrm{d}u_i}{\mathrm{d}t} = m\left\{\frac{\rho_p - \rho_f}{\rho_p}\right\}g \cdot \gamma_i + 3\pi\,\mu\, D(u_i - v_i) \quad \ldots\ldots(1A)$$

Rewriting equation (1*A*) by making the substitutions

$$U_i' = \frac{u_i}{V};\ V_i' = \frac{v_i}{V};\ t' = t \cdot \frac{V}{L}$$

we obtain:

$$\frac{\mathrm{d}U_i'}{\mathrm{d}t'} = \left\{\frac{gL}{V^2}\right\} \cdot \left\{\frac{\rho_p - \rho_f}{\rho_p}\right\} \cdot \gamma_i + \frac{3\pi\,\mu\, DL}{mV}(U_i' - V_i') \ldots(2A)$$

Now

$$3\pi\,\mu DU = m\left\{\frac{\rho_p - \rho_f}{\rho_p}\right\}g$$

$\therefore$

$$\frac{3\pi\,\mu\, DL}{mV} = \left\{\frac{\rho_p - \rho_f}{\rho_p}\right\}\frac{gL}{UV} \quad \ldots\ldots\ldots\ (3A)$$

Substituting

$$F = \frac{V^2}{Lg} \text{ and } P = \frac{UV}{gL} \text{ with } (3A) \text{ into } (2A)$$

we obtain:

$$\frac{\mathrm{d}U_i'}{\mathrm{d}t'} = \frac{1}{F}\left\{\frac{\rho_p - \rho_f}{\rho_p}\right\} \cdot (\gamma_i) + \frac{1}{P}\left\{\frac{\rho_p - \rho_f}{\rho_p}\right\}(U_i' - V_i') \ldots(4A)$$

Thus equation (3) in the original report is regained in equation (4*A*), which is identical with it when

$$\frac{\rho_f}{\rho_p} \to 0$$

For air models this ratio is normally less than 1/1000 and thus equation (3) is sufficient.

When using particles in a fluid, it is necessary to maintain equality, of the modified particle and Froude numbers between the actual system and the model; i.e.

$$P\left\{\frac{\rho_p}{\rho_p - \rho_f}\right\} \text{ and } F\left\{\frac{\rho_p}{\rho_p - \rho_f}\right\}$$

The other conditions discussed in our original report remain unaffected.

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The Trevelyan Scholarships

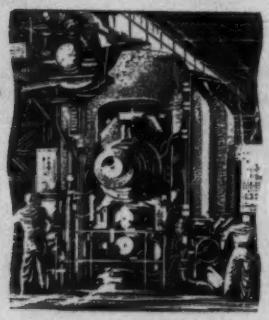
After wide consultation between a number of industrial companies and representatives of Oxford and Cambridge Universities and of public and grammar schools, the companies concerned are to provide financial support for scholarships of a new kind to enable selected boys to attend Oxford and Cambridge Universities. Dr. G. M. Trevelyan, O.M., regards the motives of the scheme favourably and has kindly agreed that the awards should bear his name.

The Trevelyan Scholarships will be open to boys of British nationality receiving full-time education at schools in the United Kingdom and will be subject to successful candidates securing admission to one of the colleges at Oxford or Cambridge. Their value will be £450 per annum for three years, extended to four years in exceptional circumstances, and will not depend on parents' income. Promises of financial support from industrial companies have already been made which will ensure that 12 to 14 scholarships can be awarded annually at each university for the initial five-year period of the scheme.

In making these awards, the sponsors believe that there are boys capable of very great contributions to the community whose capabilities cannot find full expression under the established examination system, although they possess high personal qualities and high intellectual ability. The sponsors hope that these scholarships will encourage boys to pursue a broader range of studies in the sixth form, whether related to their own special subjects or not, without impairing their chances of going to a university through lack of financial support.

With these considerations in mind, the selectors for the Trevelyan Scholarships will look for evidence that a candidate can both profit fully from an honours course and show that he has undertaken some exacting task or project. It is intended to make the first selections in November, 1958, by a committee consisting of representatives of Oxford and Cambridge Universities and industry in equal numbers under a permanent chairman.

A steering committee has been established, the members of which represent industry, schools, and the Universities concerned. The Chairman is Sir Walter Benton Jones, and Mr. R. Peddie, of 17 Westbourne Road, Sheffield 10, is secretary.



IRON AND STEELWORKS ENGINEERING

The Iron and Steel Engineers Group

Modern Developments of the Ward-Leonard Principle and Applications in Steelworks

By G. Ovens, B.Sc., M.I.E.E., and C. A. Dodd, M.B.E., A.M.I.E.E.

SYNOPSIS

The requirements of mill and processing-line drives in steelworks, where the Ward-Leonard principle finds a wide application, are reviewed.

The characteristics of the d.c. shunt motor and recent developments in design, to obtain good commutation over wide speed ranges with high outputs and under transient conditions, are briefly discussed.

A reference is made in a general way to the assessment of motor and convertor ratings. Comparisons are made between motor generator sets and mercury-arc rectifiers in a review of converting equipment.

The use of closed-loop regulating systems and the magnetic amplifier is mentioned with particular reference to tension control.

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INTRODUCTION

IN THE PAST 25 years the all-d.c. steelworks common in the early years of the century have disappeared. With the need to distribute much larger amounts of energy this was in any case inevitable but there has been at the same time a marked diminution in the use of constant-voltage d.c. motor drives, a tendency which has gone furthest on the Continent and least far in the U.S.A., with the U.K. somewhere between.

The use of a.c. for steelworks distribution has made the Ward-Leonard drive more attractive compared with constant-voltage d.c., in that a.c.–d.c. converting plant is in any case required and the cost of making this variable voltage may not be great. The main reason, however, for the increase in the use of the Ward-Leonard drive is that its characteristics are fundamentally suitable for so many steelworks applications and because these characteristics have been much improved by developments in recent years. This is a justification for a paper on the subject. The title refers to " the Ward-Leonard principle ", because this is the phrase which most steelworks electrical engineers habitually use to describe variable-voltage d.c. drives, but it is used in this paper to cover all such drives, whether the d.c. supply is obtained from d.c. generators or from rectifiers.

The paper is written entirely from the electrical viewpoint, drives having similar requirements being grouped together, although the driven plant may serve quite different purposes.

The greatly increased outputs obtainable from modern steelworks plant are, of course, due to improvements in many directions, but chiefly in the mechanical and electrical plant, where some healthy rivalry has developed. For instance, a redesigned mill puts the electrical people on their mettle to provide a drive making the most of the mill's capabilities and often they succeed so well that the mechanical plant has to be improved to stand as much caning as the ' electrics ' will take.

REQUIREMENTS AND APPLICATIONS OF DRIVES

The electric motor is used to provide mechanical power in such quantities at such time and at such speeds as best suit the operation of the driven plant. The type of motor, its electrical supply, and its electrical control are selected to meet these requirements as closely as possible consistent with the overall economics of the particular operation. Some of the great number of different operations in steelworks may best be performed with a drive using the Ward-Leonard principle.

The d.c. motor with a shunt characteristic, supplied at a controlled variable voltage (Fig. 1), has its main application in steelworks where one or more of the following requirements has to be met:

(1) Where controlled starting, stopping, or reversal is needed
(2) Where rate of acceleration or size is outside the capabilities of constant-voltage motors with armature resistance control, and where armature resistance control involves undue maintenance of heavy current contactors

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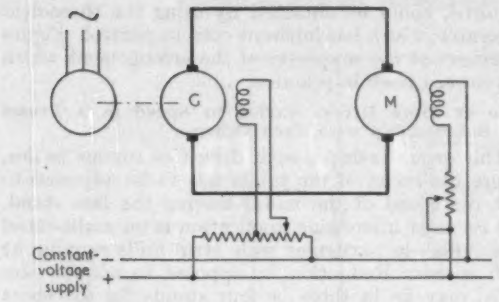


Fig. 1—Ward-Leonard method of variable speed control

(3) Where two or more drives have to be varied in speed in a preset relationship with each other
(4) Where d.c. motors are used to enable speeds to be adjusted by shunt field control, variable voltage can be used for starting and stopping a motor or a group of motors.

The four groups of requirements favouring Ward-Leonard drives can be illustrated by reviewing the various applications.

Controlled Starting, Stopping, or Reversal

The blooming or slabbing mill is the major type of plant for which the drive comes in this category. There is no need to describe the classical arrangement of a flywheel Ilgner motor-generator set, driven by a slip-ring induction motor, and a large low-speed mill motor reversing on full field but capable of field weakening for the lighter and longer passes.

The earlier installations used manual control of the generator and motor fields. Working of the mill at its maximum output without abusing the electrical plant depended on the skill of the operator. With a modern plant this is no longer true; a typical voltage control circuit[1] is shown in Fig. 2. The full forcing value of the generator exciter voltage is used by suppressing the magnetic amplifier output when the desired voltage has been reached. Similarly, the current in the motor armature circuit is limited by a feedback signal to the magnetic amplifier, and the motor is protected against stall conditions by reducing the generator voltage to a stalled value automatically. The motor field current is similarly controlled. Sequencing prevents motor field weakening before the generator voltage is above a minimum value and prevents reduction of the generator voltage unless the motor field has been strengthened sufficiently.

Fig. 3—Separate motor drive for a slabbing mill

With closed-loop rapid-response systems it is possible to use the full torque capacity of the motor and the rate of change of mill speed can be reasonably linear.

The fast response of modern slabbing mills results not only from the improvements in control devices but also from limiting the inertias of machines and gear boxes. The use of separate motors directly coupled to the top and bottom rolls means the elimination of the pinion unit with its associated losses, and the selection of long-core motors reduces the inertia.

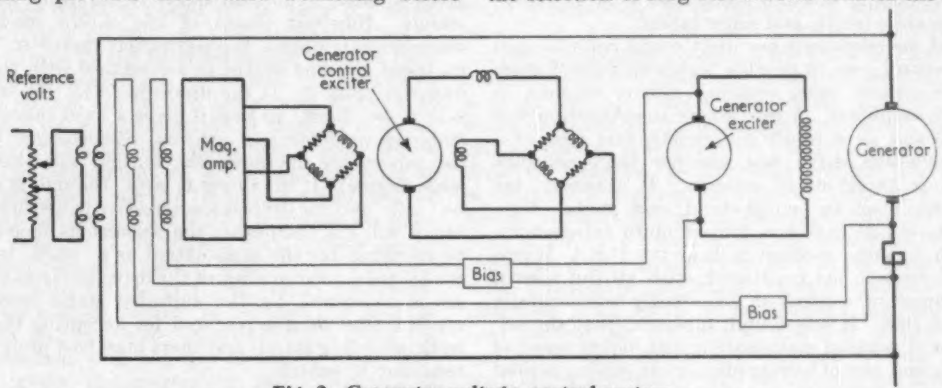


Fig. 2—Generator voltage-control system

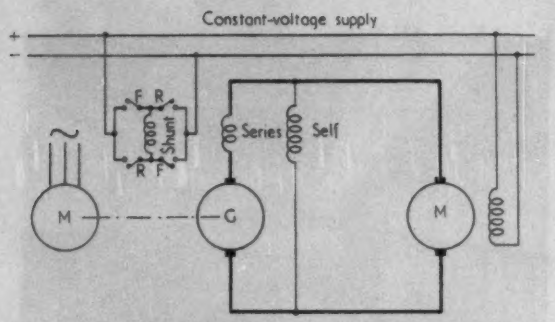


Fig. 4—Three-field generator control for screwdown gear on a multi-stand cold mill

An installation where this type of drive was used is shown in Fig. 3.

Another example of plant requiring a reversing drive is the hot or cold reversing strip mill. In this case the reversals are comparatively infrequent and the emphasis is not on the rapidity of acceleration but on the close control of tension between the reels and the mill. The changeover of duty of the reel drives from coiling to uncoiling presents problems which are satisfactorily met by Ward-Leonard systems with closed-loop control of tension by current and of motor field for progressive change of coil diameter.

Rate of Acceleration or Size outside the Capabilities of Constant-voltage Motors or their Control

The typical steelworks auxiliary drive consists of a constant-voltage d.c. motor with starting, stopping, and reversing carried out with contactors and resistances. High rates of acceleration are limited by the peak commutating current which the motor can handle. Ward-Leonard control enables this current to be maintained during the accelerating or decelerating period, whereas armature resistance control is inevitably stepped. In addition, for the same performance the heating of the motor will be less with a steady current as the r.m.s. value will be smaller. Ward-Leonard control also permits the use of much lighter contactor gear which has to handle only field currents. On plant which has to carry out a great number of operations daily the reduction in maintenance may easily offset greater capital cost, and down-time may be reduced. This is particularly true of reversing hot-mill auxiliaries such as screwdowns, manipulators, and roller tables.

Recent developments normally entail complication of the control gear to provide higher speed and more accurate control using rotating control exciters or magnetic amplifiers. A case where simplification was possible, and as a result the overall first cost of a variable-voltage drive was cheaper than constant voltage, is therefore of interest. It concerns the screwdown gear for multi-stand cold mills. Here constant-voltage gear has proved quite satisfactory, although on some modern mills in the U.S.A. Ward-Leonard control has been used, with all the normal complication of control at admittedly substantially increased cost. It was found, however, that the advantages of reduced maintenance and higher speed of response, and also of having plant more easily adapted to the use of automatic gauge control, should this be required, could be obtained by using the three-field generator, which has inherent current control. Figure 4 brings out the simplicity of the arrangement which has proved itself in practice.

Two or More Drives Varied in Speed in a Preset Relationship with Each Other

This group includes such drives as runout tables, where the speed of the tables has to be adjusted to suit the speed of the metal leaving the last stand, but its most interesting application is on multi-stand cold mills—in particular wide strip mills running at high speeds. Reduction, as opposed to temper pass mills, may be in three or four stands for the sheet gauges and normally in five stands for tinplate gauges—the thinner the final product the higher the finishing speed and the more easily broken the strip. These two factors make the control on a tinplate mill a very much more difficult proposition than on a sheet mill, a point which has been overlooked on more than one occasion in building the drives for new mills in the U.S.A., with very expensive results.

In tandem cold mills, the strip is threaded at low speed and has to be accelerated to the speed giving maximum production consistent with satisfactory running. From first principles it can be argued that the relative speeds of the stands should be maintained throughout the speed range. This, however, is not borne out in practice because of an effect known as the ' speed effect ' which is only noticeable on high-speed mills, i.e. on mills with a wide range from threading to running speed. Various theories are advanced as to the fundamental cause of the speed effect; the result of it is the equivalent of increasing the screw pressure as the speed rises. A too closely maintained ratio of stand speeds would result in the tension, particularly between the last two stands, rising excessively on deceleration, and this would almost certainly result in strip breakage. To avoid breakage it is therefore necessary to allow the individual stands to depart from the strict relationship, even though this means an increase in gauge.

On all the most recent large tinplate mills installed within the last 10–15 years, separate generators are used to supply the stand motors. The individual generator voltages are regulated so as to maintain a preset relationship which, in conjunction with the individual shunt field settings of the stand motors, gives a basic light load/speed relationship between stands. Inherent droop of the motor load/speed characteristic allows the individual motor to adjust its speed to some degree in accordance with the load imposed upon it. If the droop is small, the drive is said to be ' hard,' in that it gives a hard sensation in response to control or external disturbances (e.g. if the screws are operated, the tension will alter to a large degree). If the droop is large, the mill is said to be ' soft ' and the drive is less sensitive to disturbances, but it will not counteract the deleterious speed effect to anything like the same extent as a ' hard ' mill. It is also not a good quality of the drive as far as threading is concerned, for the unloaded stand speeds are much higher than is required for accepting the strip from preceding stands and there may be a pronounced tendency to snatch.

The Ward-Leonard system provides a ready means

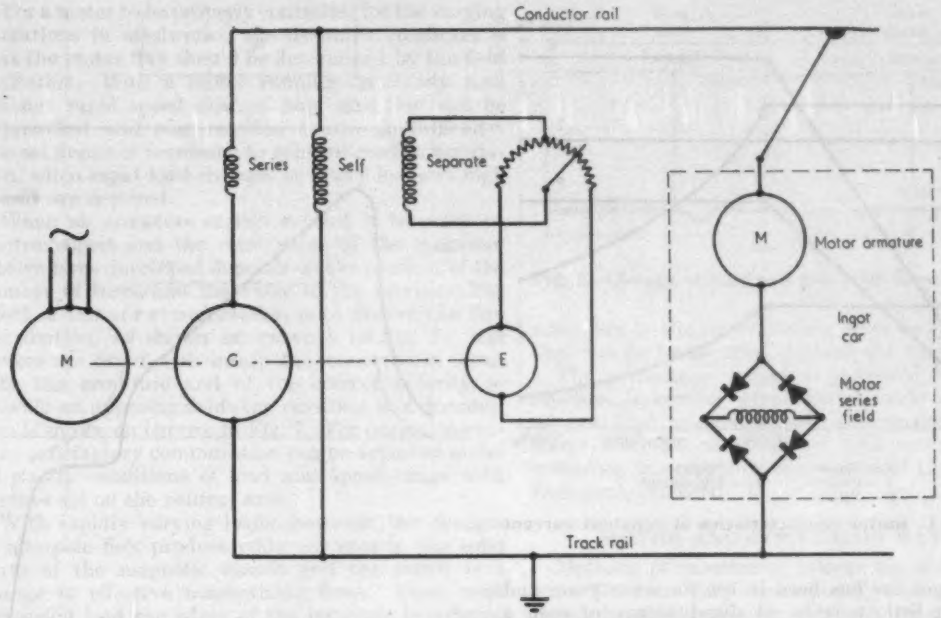


Fig. 5—Three-field generator supply to ingot car

whereby the droop of each motor can be individually and separately adjusted by exciters controlling the generator fields by a fixed yet adjustable amount. This amount is, of course, dependent on armature current but it can also be characterized in relation to mill speed so as to alter the compounding throughout the range.

In the same group can be included continuous processing-line drives where a number of motors are controlled in speed together, although subject to independent adjustment to provide such characteristics as constant tension on coiling or uncoiling, or control of a loop.

Use of Variable Voltage for Starting or Stopping a Motor or Group of Motors

There are a number of applications where variable voltage is not an inherent requirement of the drive, but where it forms a useful or, in very big sizes, possibly the only practical way of starting the plant. A typical case is on the finishing stands of a hot strip mill or on a continuous rod or bar mill. In both these cases, the motors are started before any metal is in the mill and no synchronization of starting is therefore needed. In addition, when the motors are to 'speed,' the adjustment of individual machines to suit the rolling schedule can be made on field control. A constant-voltage supply would therefore be practical. However, it is convenient and, because of the large overall capacity, usually economical to provide converting plant for the group of motors concerned. This allows voltage control to be used for starting the motors and it is also often used for holding the voltage constant with changes in supply volts, particularly where the converting plant consists of rectifiers.

The design of the drive motors for such mills has to take into account the need for close speed regulation and, as a fairly wide shunt field range is often required to meet the range of rolling schedules, this regulation has to be maintained even at weak fields.

As each slab or billet is rolled it passes into the stands successively, so that the load is suddenly applied and it is desirable to limit the impact drop as far as possible to avoid the tendency to form loops or excessive tension in the steel. Good inherent speed regulation is needed, say, between 2% and 4%, which normally raises the cost of a motor up to 20% above that with a normal regulation of perhaps between 4% and 8%. Methods of improving the steady-state regulation, such as series exciters, have very little effect on impact drop.

Ingot Car

One special case which does not fall into any of these groups may be of interest. A standard type of self-driven ingot car, with a number of collector bars on the soaking-pit side of the track and having a constant-voltage motor and resistance control, gave a great deal of trouble due to the collector rails being damaged by ingots dropping or falling against them. The car was not a very large one and it was decided, following a somewhat similar conversion elsewhere,[2] to replace the collector rails by a single trolley wire and pick-up. These supplied a series motor with a full-wave selenium rectifier connected so that the series field was always excited in the same direction. The trolley wire was fed from a variable-voltage generator using earth return through the rails. By field control, an operator at the end of the track can start, stop, and reverse the car. It was found that the single trolley wire was much less liable to damage. Should it become damaged, it could be quickly repaired.

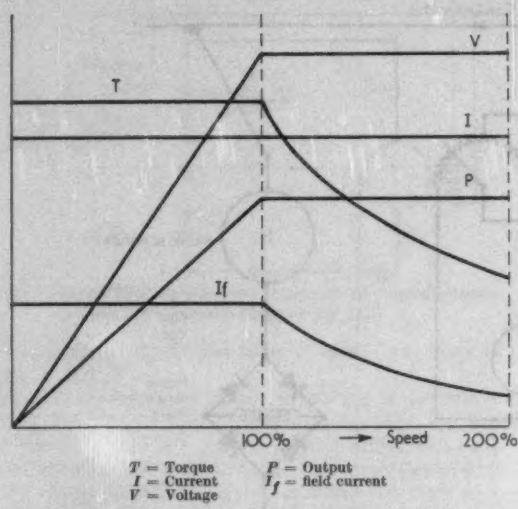


Fig. 6—D.C. motor characteristics at constant current

The ingot car has been in use for some years and has given little trouble. A disadvantage of such a simple arrangement is that end-of-track limits cannot easily be fitted, but it was felt that the driver was in no worse position than the driver of a road vehicle. Unfortunately, one day a driver set the ingot car travelling away from the mill and went off shift. The car hit a stock of cold ingots at the far end of the track and the rectifiers were knocked off. They were, however, electrically undamaged and could be refitted. The generator is of the three-field type giving automatic current limit control so that the operator can throw the controller from full speed forward to full speed reverse without trouble. A diagram of the arrangement is shown in Fig. 5. With a series motor and three-field generator, calculations should be made under all conditions to ensure that there is stability.

MOTOR CHARACTERISTICS

The d.c. shunt motor is still the most useful machine for meeting the demands of steelworks drives and the main characteristics are given in Fig. 6. Most of the applications require a reasonably flat speed/torque curve for speed matching on group drives. With field control a wide range of speed is available at constant horsepower. A shunt range of 3 : 1 is normal, and higher ranges can be obtained by compensated machines. The limit at the higher speeds with weak fields is the difficulty of achieving good commutation at heavy loads when armature reaction distorts the main field flux. At low speeds at full torque the reduced cooling means that forced ventilation is usually required. Apart from these limitations, very wide speed ranges can be obtained.

The higher speeds of response on blooming mills and the closer speed matching on tandem mills that are now required for high production have demanded further investigation in the transient response of d.c. machines.

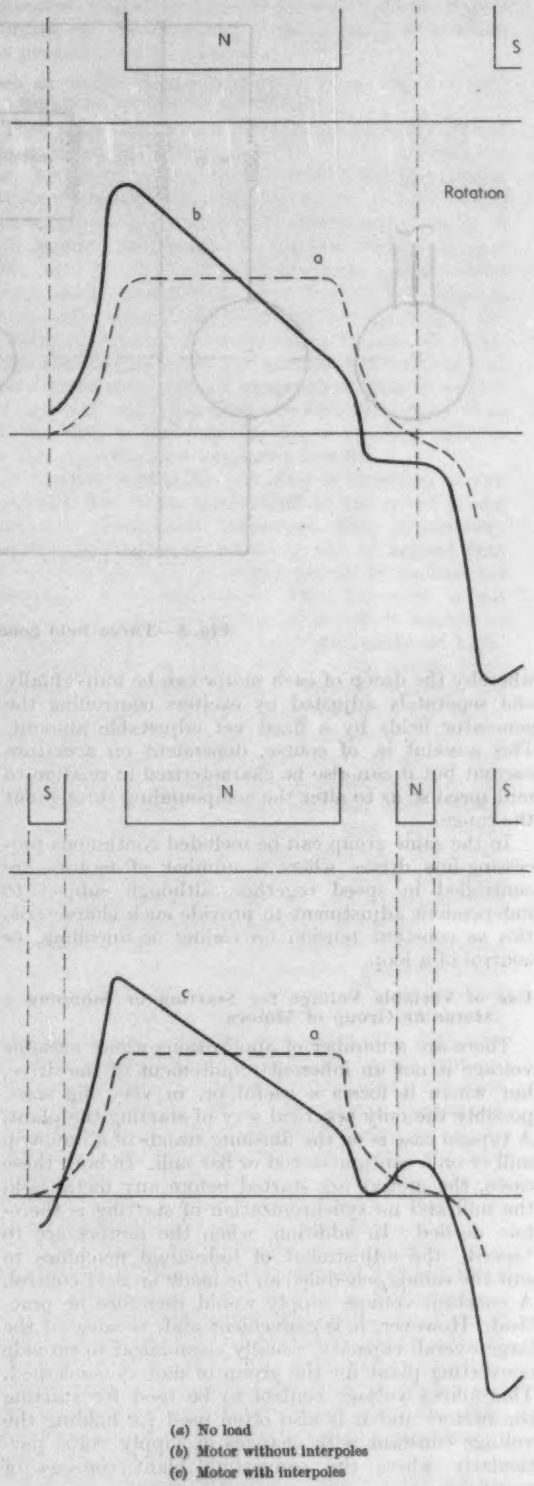


Fig. 7—Flux distribution curves for a d.c. motor

For a motor to be properly controlled for the varying operations in steelworks, the desirable condition is that the motor flux should be determined by the field excitation. With a motor running on steady load without rapid speed changes this condition can be approached and commutation causes no difficulty. Special design is necessary to achieve good commutation, when rapid load changes or heavy loads at high speeds are required.

When an armature carries current it becomes an electromagnet and the total value of the magnetomotive force developed depends on the product of the number of turns and the value of the current. The effect of the armature reaction is to distort the flux distribution, as shown on curve *b* in Fig. 7. Mill motors are fitted with interpoles connected in series with the armature and of the correct polarity to provide an opposing field; the resulting flux distribution is shown on curve *c* in Fig. 7. For normal operation, satisfactory commutation can be achieved under all steady conditions of load and speed range with brushes set on the neutral axis.

With rapidly varying loads, however, the changes in interpole flux produce eddy currents in the solid parts of the magnetic circuit and the result is a change in effective magnetizing force. Thus, with increasing load the effect of the interpole is reduced and with decreasing load the effect is increased. It is thus possible for sparking to occur, even at low currents under transient conditions; in severe cases the brushes and commutator can be damaged.

It is for this reason and to improve the overall response of the system that laminated-frame machines for heavy current reversing drives are now being manufactured. The oscillogram[1] shown in Fig. 8 was taken during a base speed to base speed test of a 12,000-h.p. slabbing-mill motor with a laminated frame and indicated the change of interpole flux with armature current.

The laminated-frame machine requires careful mechanical design to ensure rigidity and avoidance of stray magnetic loops due to pole rivets and pole liners, etc.

Interpoles have a local action in neutralizing the effect of armature reaction at the geometrical axis but do not affect the flux distribution under the main pole face. Where distortion would cause excessive voltages to be induced in the armature coils, giving too high a voltage between commutator segments, series compensating windings are installed in the pole face. Some machines have been built with these windings offset in the main pole face.[3] The compensating ampère turns are thus distributed in the field circuit in a manner similar to that of the ampère turns in the armature. Usually 75–80% compensation is provided.

Commutation on machines for these duties is still by no means an exact science; results are difficult to predict, particularly for a new design. Apart from the difficulty of assessing the correct distribution of compensating flux for the armature reactance voltage, the transfer-brush characteristics, air gaps, and the shape of the poles are very important.[4]

Another approach has been to allow additional margins for steady-load commutation, keeping the

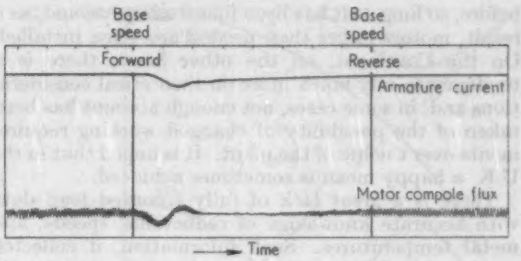


Fig. 8—Change of interpole flux with armature current

total flux in the commutating poles down to a value that can be transmitted through the surface layer.[3]

The percentage distortion produced by armature reaction is greater when heavy loads are imposed at the high speeds in the field-weakening range. Large machines are provided with cumulative compounding to strengthen the main field and reduce the high peak currents.

MOTOR AND CONVERTOR RATINGS

Methods of calculating ratings are not dealt with since, if all the data are available, these are straightforward; this section is confined to some general aspects of this important question. One has, on the one hand, to avoid the fatal mistake of putting in plant which is too small without, on the other hand, falling into the common error of calling for plant which has capacity never used throughout its life, with the additional capital and running costs that this entails.

Selection of the correct motor rating needs close co-operation between plant operating staff, the plant designer, and the electrical engineer and full understanding of one another's requirements.

An example will illustrate this need. On a strip processing line, an accelerating time of 10 s for the line to reach full speed had been specified and this entailed a motor-generator set with a motor of 675 h.p. and generator of 440 kW. Discussion showed that quality of product would not be affected if the accelerating time were increased and, as each full coil would take 20 min to process, output would not be appreciably reduced. The lower peak torque capacity needed with the longer accelerating time allowed the generator size to be reduced by 140 kW and the motor by 210 h.p.

Ideally, one might argue that the operating staff should provide a complete set of operating data (e.g. rolling schedules for a mill). The mechanical people should convert these to speeds and torques and the electrical people should calculate the ratings. This would make it easy for the electrical engineers, but it rarely works out in this way. Assessment of ratings is almost more of an art than a science, being based very much on past experience. As a result, mill plant of the same basic design and used to produce the same product may in different places have differently rated drives.

In the U.S.A., in spite of the emphasis on research and progress, there is, in this particular field, a strong tendency to repeat uncritically what has been done

before, so long as it has been found adequate and, as a result, motors larger than needed are often installed. On the Continent, on the other hand, there is a tendency to rely much more on theoretical considerations and, in some cases, not enough account has been taken of the possibility of changed working requirements over the life of the plant. It is hoped that in the U.K. a happy mean is sometimes achieved.

There is a great lack of fully recorded load data with accurate knowledge of reductions, speeds, and metal temperatures. Such information, if collected over a wide field, would be a great help in correctly assessing ratings. Perhaps the British Iron and Steel Research Association might do even more in this field.

The probability that, if the plant is capable of greater output, it will in due course be used to produce it should be taken into account in assessing ratings and here, for an economical overall design, one must ensure that the capabilities of the mechanical and electrical plant are on a par.

The rating of a d.c. motor, whether driving a mill or auxiliary plant in a steelworks, is determined firstly by the peak torque requirements (and here safe commutation is usually the limit) and secondly by the r.m.s. or heat rating. Sometimes one and sometimes the other fixes the size of the motor.

Where geared drives are used, the actual speed can be selected to suit the design of the motor, but with direct drives too high a speed for the horsepower required may entail double-armature machines. The shunt field range needed affects the motor size.

On reversing drives, the torques required for acceleration and deceleration may form a large, or even a dominant, proportion of the total motor load, and a major part of this torque may be required to accelerate the motor armature. Where there are gears, accelerating torques can be reduced by keeping the speed of the motor low. Alternatively, double-armature machines, which may be needed in very big sizes where the speed is too high for a single machine, also reduce inertia, since they are equivalent to a longer and smaller-diameter motor.

Whereas motor ratings must be decided in relation to what they drive and not, therefore, primarily by the electrical engineer, the rating of the converter supplying it is entirely within his competence.

Where a generator or group of generators supply a motor or motors on a single piece of plant, generator ratings will normally be the same as the motor ratings. Where, however, generators supply a number of motor drives, such as a multi-stand mill or processing line with a common busbar arrangement, some diversity is often allowable since there may be no rolling schedule in which all stands are on full load.

Similar considerations apply when rectifiers are used. An illustration of the possible saving was recently shown in considering the six finishing stands of a wide strip mill, where the alternatives were separate rectifiers for each of the six d.c. motors or a common busbar arrangement. The separate-rectifier scheme has advantages such as lower fault capacity, but on the rolling schedules expected it was possible to apply a diversity on the common busbar scheme, allowing a saving in installed rectifier capacity of 2500 kW.

The advantages of diversity can often be even greater on driving motors of motor-generator sets where a number of generators are driven by a single motor, as on multi-stand mills and strip processing lines. On a recent cold-mill installation the driving motors of the two motor-generating sets were each reduced from 11,500 (the size which had been installed previously) to 9000 h.p., with a saving of some £100,000 in capital cost and at least £3000 a year in running costs.

The possibility on blooming and slabbing mill drives of taking advantage of the high capacity of modern power systems to eliminate the variable-speed Ilgner set is mentioned elsewhere in this paper. It has the effect of increasing the size of the driving motor of the motor-generator set, since on a constant-speed set this motor must be capable of supplying the peak mill motor torque. Rectifiers would, of course, have to be rated on the same basis.

CONVERTING EQUIPMENT

A Ward-Leonard or variable-voltage d.c. drive is classically supplied from a variable-voltage d.c. generator, normally forming part of a motor-generator set with an a.c. driving motor. There is, of course, no reason why the generator cannot be driven from any other source of mechanical power and, for large sets of the Ilgner type, where feed-back energy can be absorbed in accelerating a flywheel, it is sometimes possible to show, on paper, large savings by using a steam- or even gas-turbine drive. However, lack of flexibility and the disadvantage of bringing another form of power into the motor room are usually and rightly considered overriding objections.

The generator set, at any rate until quite recently, has maintained its place for the following reasons:

(1) The ease of control and reversal of voltage
(2) The automatic feedback of energy into the supply system
(3) The possibility of using an Ilgner set with a flywheel to smooth out the heavy load fluctuations produced by large reversing mills. Most steelworks are now interconnected with large modern power systems, where the peak loading is of little importance and the use of an Ilgner set with its greater cost and higher losses may not be justified
(4) Where a synchronous driving motor is used, power factor can be controlled and correction provided.

The disadvantages of the motor-generator set are:

(1) The high light-load losses. Even on very heavily worked mills, the number of hours a year during which the motor-generator set is running light or with only a small load is high.
(2) Heavy foundations are required and installation costs are high.
(3) Maintenance, including that of brush gear and commutators, is appreciable.

However, the performance of modern reversing mills using motor-generator sets has set a high standard for any alternative equipment. On large slabbing mills, for instance, reversal times from base speed in one direction to base speed in the other of less than 1 s have been achieved.

The present rival to the motor-generator set is, of course, the mercury-arc rectifier and the rest of this section is devoted to its application to variable-speed drives.

In recent years, development of boundary layer

rectifiers has been very rapid, and equipment of this type will no doubt be used to provide a variable-voltage d.c. source either in conjunction with induction regulators or saturable reactors in the a.c. side or, looking further ahead, possibly with some form of 'grid' control. Of those at present being manufactured, the silicon rectifier seems to be the best adapted for this duty.

The advantages of a mercury-arc rectifier in providing a variable-voltage supply are:

(1) Its higher efficiency, particularly at low loads. Since the arc drop in a rectifier is a big proportion of the total voltage and is nearly an absolutely constant value, then the higher the design voltage, the higher the efficiency. It is therefore desirable to use as high a voltage as possible. A compromise has to be found which does not prejudice the design of the d.c. motor. Below about 250 V, the motor-generator set becomes more efficient.
(2) The rectifier is static plant needing only light foundations and is comparatively easy to install.
(3) Unlike a synchronously driven motor-generator set, it does not raise the fault level of the a.c. system.
(4) Its speed of response is very fast. This makes stabilizing of voltage-control systems more simple and enables a rapid change in voltage to be accurately controlled where needed.

The disadvantages of a rectifier are:

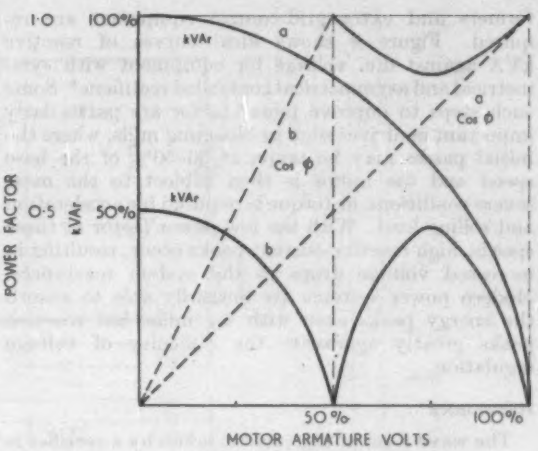
(1) It is essentially a piece of apparatus providing constant voltage; to produce variable voltage means additional complication, nearly always the introduction of grid control. This has the secondary disadvantage that the power factor deteriorates with reduction in voltage. With plant which normally runs at substantially full voltage and where reduced voltages are only required for starting, this objection is not serious and rectifiers are normally used.
(2) A further complication is needed to obtain reversal of voltage or regeneration of power into the power system.
(3) Rectifiers produce harmonic voltages which can cause trouble in the supply system.

It is of interest to deal with these last two points in more detail.

Reversal of Power Flow

The mercury-arc rectifier can conduct current only in one direction—from anode to cathode—but a rectifier with grid control can be arranged not only to control the voltage, but also to operate as an inverter, accepting power from the d.c. side and returning it to the a.c. system. There are three methods of achieving this:

(1) *By installing separate cross-connected forward and reverse power rectifiers.* The objection to this method is the high cost, but where the amount of reverse power is only a small proportion of the forward power, such as in a unidirectional mill where power has to be absorbed only in stopping, it may be economical to install a small reverse rectifier.
(2) *By using reversing switches for the motor armature.* This method is fast, but entails providing very heavy reversing contactors for the larger equipments. They do not have to make or break current, as the switches are operated at zero current, but up to say 600 reversals per hour may be required. Switches for manufacture in the U.K. would require special development. Some have, however, been developed in Germany.
(3) *By using reversing switches or cross-connected rectifiers in the motor field.* The field currents are, of course, much smaller than the armature currents but, on the other hand, the inherent time constant of the motor field means that the total time of reversal is increased, even if a large measure of field forcing is used.



(a) With symmetrical grid control
(b) With asymmetrical grid control

Fig. 9—Reactive power curves for symmetrically and asymmetrically controlled mercury-arc rectifiers

Rectifiers have not been applied to reversing mill drives in the U.K., although no doubt they would be carefully considered for any new installation, probably in conjunction with armature reversing switches. A rectifier-fed drive for a 300-h.p. mine winder was installed in 1955.[5] Field reversing switches were used, but it appears that armature switches would have been preferred if suitable ones had been available.

The use of rectifiers for mill drives has developed further on the Continent than in the U.K., and in Germany it is present policy to install rectifiers on all mill drives where the supply system can tolerate the high peak loads. Blooming mills with main motors totalling over 9000 h.p. have been installed and are apparently working well. Furthermore, a four-stand cold-reduction tandem mill supplied by rectifiers has been running for about a year.

Power Factor

The power factor of the supply taken by a rectifier is reduced in proportion to the reduction in voltage by grid control. Normally a margin has to be allowed for voltage control at the top of the range, and some reduction in power factor results, even when the drive is running at rated voltage. For drives which run mainly at full voltage, such as the finishing train in a hot strip mill, the reduction in power factor is not serious.

Where greater voltage reduction is required, power factor can be improved by asymmetrical control of the rectifier grids, a method used on the Continent. In this method some of the anodes are controlled so that they are working at nearly full voltage while grid suppression and variation of the pulsing instant is used on the remaining anodes. Interphase trans-

formers and extra grid-control equipment are required. Figure 9 shows ideal curves of reactive kVA against d.c. voltage for equipment with symmetrical and asymmetrical controlled rectifiers.[6] Some such steps to improve power factor are particularly important on drives such as blooming mills, where the initial passes may be taken at 30–50% of the base speed and the motor is then subject to the most severe conditions, as torque is required for acceleration and rolling load. With the low power factor at these speeds, high reactive-current peaks occur, resulting in increased voltage drops in the system reactances. Modern power systems are normally able to absorb the energy peaks even with big mills, but reactive peaks greatly aggravate the difficulty of voltage regulation.

Harmonics

The wave form of the current taken by a rectifier is not sinusoidal, but has a mixed harmonic content, the amplitude of the harmonics decreasing with the number of phases.

For a rectifier with a given number of phases, the harmonic currents are proportional to the rectifier loading and the voltage distortion caused by these currents is inversely proportional to the capacity of the system (i.e. to its short-circuit MVA).

The harmonic voltages can cause interference with other equipment or, in an interconnected system, with that of other consumers. In a recommendation of the Central Electricity Authority, details are given of the ratio of connected rectifier kilowatts to short-circuit MVA with maximum harmonic current permissible as a percentage of the rated current.

In large rectifier installations in this country, the number of phases chosen has usually been 24 or more, and the system short-circuit MVA has been high. The distortion produced has not caused serious interference, but each installation requires individual consideration.

Where abnormally large rectifiers are connected electrically close to a combination of inductive and capacitative components, such as transformers and power-factor correcting capacitors, serious voltage disturbance can be caused owing to resonance at harmonic frequencies. This can be overcome by filtering out the resonant frequency, or detuning the resonant circuit. Some trouble has been experienced with Petersen earthing coils on overhead distribution systems and occasionally on induction-type meters.

Telephone interference in steelworks is not likely to be serious unless the telephone cables are run close to the rectifier busbars.

The proportion of mercury-arc rectifier capacity to total installed capacity of steelworks is tending to rise. Whilst it is generally possible by careful design of new steelworks plant to overcome interference from the rectifiers, it is important that too high a standard of freedom from interference should not be set. A high standard of freedom from distortion in the power supply is costly to attain and may tip the balance against the use of rectifiers.

In Germany it seems that the regulations are less stringent and many rectifier equipments have been concentrated in one area of the country. At one works, all main drives are supplied from rectifiers, and the slabbing mill now in course of erection will also have mercury-arc rectifier equipment. It has been noted that, although the harmonic distortion decreases with the increasing number of phases, the present tendency in Germany is to decrease the number of phases to twelve or even six.

Future Developments

It is certain that mercury-arc rectifiers will be widely used in the U.K. on large variable-speed applications, where practice has hitherto been to use motor-generator sets. Where large powers are required, the increased overall efficiency enables substantial savings to be made in running costs, possibly as much as 10% on a blooming mill and at least 5% on a tandem cold mill.

CONTROL

The increased speed of modern plant with the higher quality of product required has resulted in an increase both in the amount and in the complexity of the control equipment. Closed-loop regulating systems have been essential to achieve a fast response, and it is mainly the increased knowledge about the behaviour of these systems that has made the improved performances possible. The Ward-Leonard system is particularly well suited to the application of closed-loop regulating techniques. The Ward-Leonard d.c. generators and motors are in themselves amplifiers in that a large quantity, the armature output, can be controlled by a small quantity, the field. Control systems for speed, voltage, current, torque, and position find many applications in the steel industry but it is not possible within the scope of this paper to deal with many of these, however interesting they may be. Control of strip tension is taken as an example, for this involves in many cases the application of closed-loop regulating systems to both the Ward-Leonard generator and the motor.

Tension Control

In hot and cold reduction mills the tension is usually set manually by the initial field settings and maintained to a degree by the inherent characteristics and trimmed manually during operation.

On coilers and on modern process lines, control of the individual motors is necessary to maintain tension in the strip.

Tension Control by Current

An interesting example of control applied to Ward-Leonard drives occurs in the equipment for driving a centre-driven coiler or uncoiler for strip. In many applications it is necessary to coil or pay off under steady tension conditions. The tension requires to be regulated between close limits both during steady running and during speed changes. Sometimes a light tension is required and, if this requirement is not accurately met by the control system, there may be at times a tendency to throw slack.

With a Ward-Leonard system the main generator voltage is varied in approximate proportion to strip speed. In the simple case this voltage is applied to the coiler or uncoiler motor and the motor field is regulated to maintain the motor armature current at some fixed value to give the desired tension. Provided

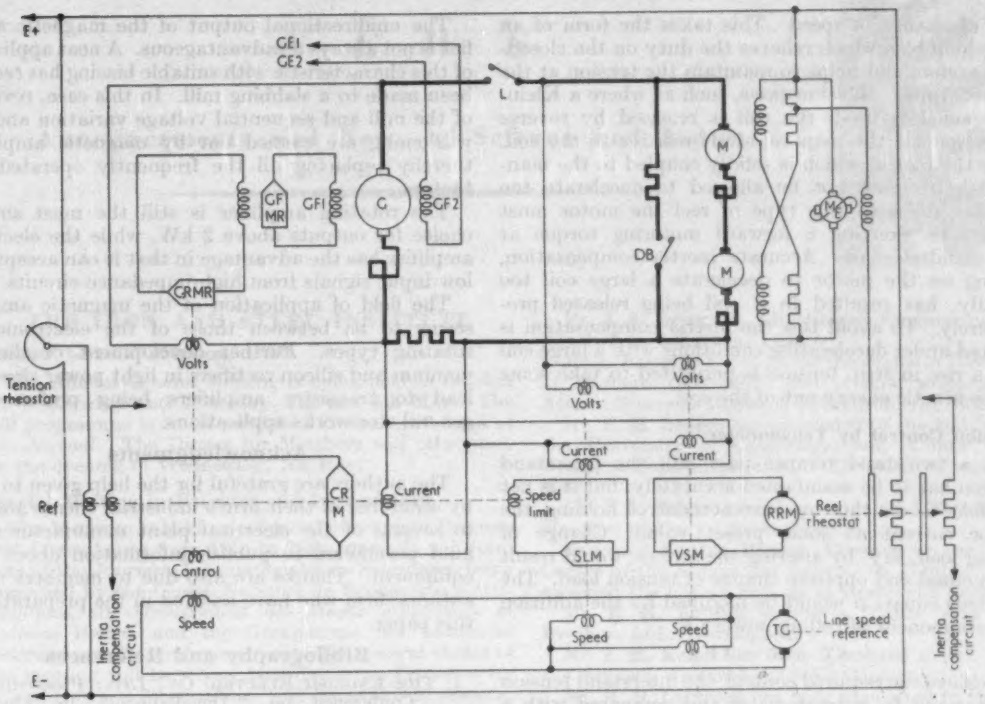


Fig. 10—Tension reel control

that this is so, the strip speed may be varied by generator voltage control and the tension will remain substantially constant because the following fundamental relationship holds:

Power in strip	= coiler motor power
Speed × pull	= electrical input h.p. × efficiency (including losses, e.g. friction in coiler)
S (ft/min) × T (lb)	= volts × amps × 0·746 × efficiency
or ST	$= XVA$

If thus the motor voltage (V) is proportional to strip speed (S), the current regulated at some preset value will give a constant tension. This implies a constant value of the factor X which is not so in practice, for it has components dependent on speed, on coil size, and on the variation of motor efficiency throughout the whole range.

Where a separate coiler generator or booster is used the above relationship holds good, provided that the motor terminal voltage or, more precisely, the motor internal e.m.f. is maintained proportional to strip speed.

In order to give accurate control of tension with quick response throughout the whole range of strip speed and coil size it is usual to regulate the current by generator or booster field control using a rapid-acting regulator for this purpose. The motor terminal voltage is regulated by adjusting the motor field by a comparatively slow-acting regulator which, in effect, follows up the change of coil diameter by adjusting the field by a corresponding amount.

Figure 10 shows the basic circuits for controlling a coiler motor for a cold strip mill.[7] As the coil diameter increases the motor slows down, thereby reducing its e.m.f. and increasing the armature current. The current regulating magnetic amplifier *CRM* compares the armature current signal with a set signal corresponding to the set tension. An increase in armature current causes the generator value to be reduced by means of magnetic amplifiers *GRMR* and *GFMR*. The change of generator voltage, while reducing the armature current, upsets the basic volts/speed relationship for the motor, and this is restored by strengthening the motor field by means of magnetic amplifier *VSM* and a rotating amplifier. Although not shown on the diagram, suitable feedback circuits are incorporated to provide stability.

If no change of diameter is involved, as in the control of tension by bridle rolls, the above conditions are simplified and tension control is accurately achieved by current control alone.

In both cases (i.e. ranging and fixed diameters), steps are taken either to limit the speed when there is no strip to constrain the roll to run at the correct speed by speed limit circuits, or to make the tension control dependent on manual application after threading.

The current and voltage regulators are independent closed-loop servo-control systems, i.e. they are error-actuated. It is a characteristic of such systems that deviation from the set value of tension, for instance, is increased under transient conditions during change of speed. Compensation for this increased error is applied in value dependent on the inertia and

rate of change of speed. This takes the form of an open-loop bias which relieves the duty on the closed-loop system and helps to maintain the tension at the desired value. In some cases, such as where a Klein-type coiler is used, the coil is released by reverse movement of the mandrel shaft relative to the coil. Here the motor, which is solidly coupled to the mandrel shaft, must not be allowed to decelerate too rapidly, for with this type of reel the motor must always be exerting a forward motoring torque at the mandrel shaft. Accurate inertia compensation, calling on the motor to decelerate a large coil too rapidly, has resulted in a coil being released prematurely. To avoid this the inertia compensation is relaxed under decelerating conditions with a large coil and a rise in strip tension is permitted to take some of the kinetic energy out of the coil.

Tension Control by Tensiometer

In a two-stand temper pass mill the interstand tension has to be maintained accurately, but it is not possible to do this by current control holding the motor current at some preset value. Change of rolling load, say, by altering the screws, would result in an equal and opposite change of tension load. The previous equation would be modified by the addition of a component of rolling power (R)

$$ST + R = XVA$$

To achieve the required control, the interstand tension is measured by a tensiometer and compared with a desired value, the motor current being adjusted to correct for any deviation.

Magnetic Amplifiers

As mentioned previously, the generators and motors of the Ward-Leonard system are in themselves amplifiers and are suitable as components in closed-loop servo systems. For high accuracy control, however, the comparison between a measured value and a desired value is carried out at low power levels and further amplifying components are necessary to provide sufficient power to supply the generator or motor fields.

In recent years much development has been carried out on magnetic amplifiers, whose greatest merit lies in the fact that they are entirely static and require less maintenance than rotating amplifiers. In addition, where many magnetic amplifiers of a similar type are used in one installation they can be built in tray form for easy replacement. In many cases they will operate satisfactorily off a 50-c/s supply, but where a rapid response is required, or where the bulk or weight of equipment has to be kept small, a higher-frequency supply of 400 c/s is necessary.

One characteristic of magnetic amplifiers is that the output is unidirectional. This means that a single magnetic amplifier cannot give an output varying continuously from a maximum value in one direction through zero to a maximum value in the opposite direction. If such an output is required—and this is usual in many servo applications—it can be done either by biasing (which introduces other difficulties) or by connecting the input fields of two identical magnetic amplifiers in series with the outputs in push-pull.

The unidirectional output of the magnetic amplifier is not always disadvantageous. A neat application of this characteristic with suitable biasing has recently been made to a slabbing mill. In this case, reversing of the mill and sequential voltage variation and field weakening are carried out by magnetic amplifiers, thereby replacing all the frequently operated contactors.

The rotating amplifier is still the most suitable choice for outputs above 2 kW, while the electronic amplifier has the advantage in that it can accept very low input signals from high-impedance circuits.

The field of application of the magnetic amplifier seems to be between those of the electronic and rotating types. Further development of the germanium and silicon rectifiers in light power sizes may lead to transistor amplifiers being preferred for general steelworks applications.

Acknowledgments

The authors are grateful for the help given to them by a number of their firm's industrial clients and also to several of the electrical plant manufacturers for kind permission to publish information about their equipment. Thanks are also due to members of the authors' firm who have assisted in the preparation of this paper.

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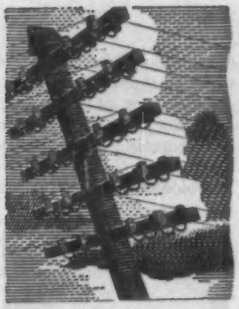
N E W S

Announcements and News of Science and Industry



THE IRON AND STEEL INSTITUTE

Annual General Meeting, 1958

The Annual General Meeting is being held in London on Wednesday and Thursday, **7th and 8th May.** The full programme is to be published in the April issue of the *Journal*. The Dinner for Members will take place on the evening of Wednesday, 7th May.

Special Meeting in Belgium and Luxembourg

The Institute is holding a Special Meeting in Belgium and Luxembourg from **18th to 28th June.** It will form part of the Journées Internationales de Sidérurgie, being organized by the Centre National de Recherches Métallurgiques, the Groupement des Hauts Fourneaux et Aciéries Belges, and the Groupement des Industries Sidérurgiques Luxembourgeoises. The general theme of the meeting will be "New Manufacturing Processes for Use in Iron and Steelmaking"; at the technical sessions, papers from Austria, Belgium, Canada, France, Germany, Italy, Luxembourg, Sweden, the U.K., and the U.S.A. are to be presented and discussed. The meeting will be divided into three sections, located at Liège, Luxembourg, and Charleroi and Brussels. A full programme of works visits and social functions has been arranged, in addition to the technical discussions, and there is also a ladies' programme.

The outline programme was published in the December, 1957 issue (p. 344); the full programme will appear shortly in the *Journal*.

NEWS OF MEMBERS

Mr. **A. Atha** has been appointed a Managing Director of Guest, Keen and Nettlefolds (Midlands) Ltd.

Mr. **K. T. Brightman** has been appointed Assistant Technical Sales Manager of Davy British Oxygen Ltd.

Mr. **D. L. Campbell** has been appointed Managing Director (Sales) of the Metallurgical Equipment Export Co., Ltd.

Mr. **P. M. Cook** has been elected a Fellow of The Institution of Metallurgists.

Mr. **S. J. Cort** has retired from the position of Vice-President in charge of steel operations of the Bethlehem Steel Company, after 40 years' service.

Mr. **A. G. Davies** has transferred to the Stourvale Works of Richard Thomas and Baldwins Ltd., Kidderminster, following the closing-down of the Clayton Works, Pontardulais.

Mr. **P. R. Dhar** is now Head of the Department of Metallurgical Engineering at the Indian Institute of Technology, Kharagpur.

Mr. **M. F. Dowding** has been appointed Chief Engineer of the Davy and United Engineering Co., Ltd., in succession to the late Mr. W. W. Franklin.

The Rt. Hon. **the Earl of Dudley** is retiring from the Board of the Steel Company of Wales Ltd.

Mr. **H. H. Egginton** is now Works Manager of Teddington Refrigeration Controls Ltd.

Mr. **J. S. Exley** has left Mitchell Engineering Ltd. to join B.T.R. Industries Ltd.

Mr. **L. Fairman** has obtained the degree of B.Sc. (Hons.) at Durham University, and is now working at the Atomic Weapons Research Establishment, Aldermaston.

Mr. **P. M. George** has proceeded to the B.A. degree of the University of Cambridge, and is now with the Research Department of I.C.I. Metals Division, Birmingham.

Mr. **J. L. Harrison** has been appointed Technical Sales Manager of Davy British Oxygen Ltd.

Monsieur **Eugène Herzog,** Director of the Research Centre of the Société des Aciéries de Pompey, has been awarded the Grand Prix de la Recherche Scientifique.

Mr. **B. K. Kaplish** is now a trainee with Rylands Brothers Ltd., Warrington.

Mr. **I. M. Kemp** has been appointed to the Board of the Appleby-Frodingham Steel Company.

Mr. **K. G. Lampson** has been appointed to the Boards of the Appleby-Frodingham Steel Company, Steel, Peech and Tozer, Samuel Fox and Co., Ltd., and the Workington Iron and Steel Company.

Mr. **Herbert Meier** has rejoined the Georg Fischer Company, Schaffhausen, Switzerland.

Mr. **J. F. Perry** has retired from the Board of Metropolitan-Vickers Electrical Co., Ltd.

Lieut.-Cdr. **G. W. Wells** has been appointed to the Boards of Samuel Fox and Co., Ltd., the Workington Iron and Steel Company, and Steel, Peech and Tozer.

Mr. **D. R. Wood** is now a member of the development staff of the D. and R. Department of The Mond Nickel Co., Ltd.

Obituary

Mr. **William Geary,** F.I.M. (*elected* 1935; *Williams Prizeman*, 1937), of Scunthorpe, on 12th January, 1958.

Dr. **R. H. Greaves,** F.I.M. (*elected* 1920), of London, on 20th January, 1958.

Mr. **Job Holland,** O.B.E. (*Member* 1904–1948), of Jersey, C.I., on 28th December, 1957.

Senor Don **Emilio Ruiz Gonzales** (*elected* 1946), of Guipuzcoa, Spain, on 3rd December, 1957.

Prof. Dr. Eng. **Tibor Vér** (*elected* 1955), formerly of Austria, in the U.S.A. on 30th November, 1957.

Mr. **Percy William Waugh** (*elected* 1953), of Dartford, on 20th November, 1957.

CONTRIBUTORS TO THE JOURNAL

P. D. Blake, M.A. (CANTAB.)—Research Physical Chemist, Fundamental Research Team, Murex Welding Processes Ltd.

Mr. Blake graduated in 1944 from Christ's College, Cambridge, with B.A.Hons. (Natural Sciences Tripos), and obtained his M.A. four years later. He was a member of the Physical Chemistry Research Section of the Metropolitan-Vickers Electrical Co., Ltd., from 1944 to 1947, and carried out basic research on welding electrodes. He joined Radiation Ltd. in 1947 as a research physical

chemist, where he was engaged on basic research into vitreous enamelling. From 1951 to 1952 Mr. Blake worked on Ministry of Supply projects at Plessey Ltd., and in 1951 he joined his present company, where he has been investigating gases in metals. He is a member of the Group on Gases and Non-Metallics of the B.I.S.R.A. Physical Chemistry of Steelmaking Committee.

P. D. Blake

G. Ovens

C. A. Dodd, M.B.E., A.M.I.E.E.—On the staff of McLellan and Partners.

Mr. Dodd was educated at Faraday House and served his apprenticeship with the English Electric Co., Ltd., at Rugby and Stafford. He served in the Royal Naval Volunteer Reserve from 1943 to 1946 and then rejoined the English Electric Co., Ltd., where he worked on the application of electric drives for steelworks' use. He joined his present firm in 1954.

S. Ignatowicz, DIPL.ING.—Physical Chemistry Section, British Iron and Steel Research Association.

Mr. Ignatowicz was born and educated in Poland. His studies in chemistry at the Technical University in Lwow were interrupted by the war, which he spent in the U.K. as a member of the Polish Air Force. Mr. Ignatowicz completed his studies at the Polish University College in London after the war, and carried out postgraduate research in high-temperature electrochemistry and thermodynamics at Imperial College. He joined B.I.S.R.A. in 1950.

G. Lagerberg

C. A. Dodd

G. Lagerberg, S.M.—AB Atomenergi, Stockholm.

Göran Lagerberg was born in Stockholm in 1925, and studied at the Royal Institute of Technology there, graduating in 1951. After graduating he stayed on for six months as a research assistant in physical metallurgy, and then joined the Research Department of Stora Kopparbergs Bergslags AB, Domnarvet. He spent one year's leave during 1954 and 1955 at the Massachusetts Institute of Technology, obtaining the degree of Master of Science in metallurgy. Mr. Lagerberg, whose work has been chiefly devoted to the physical metallurgy of mild steel, left Domnarvet in February of this year to join AB Atomenergi, Stockholm.

G. Ovens, B.SC., M.I.E.E.—Technical Director, McLellan and Partners.

George Ovens was educated at Daniel Stewart's College and Edinburgh University, where he graduated with First Class Honours in 1931. After a student apprenticeship with The British Thomson-Houston Co., Ltd., at Rugby he joined the firm's Industrial Engineering Department, where he dealt with sectional drives for paper-making machines and, during the war, with servomechanisms for radar. In 1945 he took up an appointment with McLellan and Partners, with whom he has been closely associated on developments in the tin-plate industry of South Wales. He was made a Technical Director of the firm in 1954.

J. D. Stringer, B.SC.(TECH.)—Mechanical Working Division, British Iron and Steel Research Association.

Mr. Stringer was educated at King Edward VII School, Sheffield. After service with the Royal Engineers, followed by a short engineering apprenticeship, he entered Manchester University (Faculty of Technology) and obtained a B.Sc.(Hons.) degree in mechanical engineering in 1951. After graduating he worked at The Pressed Steel Co., Ltd., on engineering and heat-transfer problems in refrigeration. He joined B.I.S.R.A. in 1953 and has worked on engineering, heating, deformation, and operational problems in forging. Since 1955 he has been in charge of forging research.

J. D. Stringer

A. Tomlinson

A. Tomlinson, A.MET., A.I.M.—Mechanical Working Division, British Iron and Steel Research Association.

Mr. Tomlinson is a native of Sheffield and was educated at Carfield School. He received his initial metallurgical training with the Brown-Firth Research Laboratories and obtained an Associateship in Metallurgy at Sheffield University in 1948. From then until 1955 he worked on metallurgical problems associated with the production of alloy-steel forgings and rolled products at Thomas Firth and John Brown Ltd. In 1955 he joined the B.I.S.R.A. Mechanical Working Division to work on forging research.

IRON AND STEEL ENGINEERS GROUP

Thirty-sixth Meeting

The Thirty-sixth Meeting of the Iron and Steel Engineers Group will be held at the offices of the Institute, 4 Grosvenor Gardens, London, S.W.1, on Wednesday and Thursday, **16th and 17th April,** from 10 A.M. to 4 P.M. each day.

The first day will be devoted to the discussion of a paper on "Modern Developments of the Ward-Leonard Principle and Applications in Steelworks," by G. Ovens and C. A. Dodd (McLellan and Partners), in the morning and to a general discussion of "Electrical Maintenance in Iron and Steel Works" in the afternoon.

The second day will be devoted to the discussion of papers from B.I.S.R.A. on "The Application of Remote Position Control to Screwdown Drives of Rolling Mills" by L. N. Bramley, G. K. Steel, and F. Seredynski, and "The Application of Remote Position Control to Forging" by F. Seredynski, G. K. Steel and others, in the morning, and "Translator Control" by W. N. Jenkins in the afternoon.

BRITISH NUCLEAR ENERGY CONFERENCE

Forthcoming Meetings

The following meetings have been announced:

Tuesday, 8th March, 5.30 p.m.

"Further Developments in Nuclear Power" (*James Forrest Lecture*), by Sir John Cockcroft—Institution of Civil Engineers, 1–7 Great George Street, London, S.W.1.

Wednesday, 16th April, 5.30 p.m.

"Operational Experience—Calder Hall," by K. L. Stretch—Institution of Electrical Engineers, Savoy Place, London, W.C.2.

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

Staff Changes

Mr. K. P. Perry, formerly Assistant Technical Secretary of the Steelmaking Division, is now Technical Secretary of the Mechanical Working Division. He succeeds Mr. O. H. Kemmis, who has joined the Development and Steel Users Section. Miss I. Richert and Mr. V. H. Guy have been appointed Assistant Technical Secretaries of the Mechanical Working and Steelmaking Divisions respectively.

EDUCATION

Foundry Sands Fellowship

The South African Branch of The Institute of British Foundrymen has established a Foundry Sands Fellowship with the South African Council for Scientific and Industrial Research. Applications are invited from holders of B.Sc. degrees or the equivalent. Appointment will be for three years in the first instance; the basic initial salary will be in the range £510–1380 p.a., with £234 cost-of-living allowance for married men.

Applications giving age, qualifications, and experience should reach the Acting Director, National Mechanical Engineering Research Institute, P.O. Box 395, Pretoria, South Africa, before 31st March, 1958.

Mond Nickel Fellowships

The Mond Nickel Fellowships Committee announced recently the award of Fellowships for 1957 to the following applicants:

Mr. E. J. Williams (John Summers and Sons, Ltd.), to study the research and production techniques, quality control, and metallurgy of high-quality low-carbon strip steel in the U.K., the Continent, and the U.S.A.

Mr. R. J. D. Acheson (Mufulira Copper Mines Ltd.), to study the extraction and refining processes for copper and allied metals, with particular reference to economic aspects.

The Mond Nickel Fellowships Committee now invites applications for Fellowships of an approximate value of £900–1200 for 1958. Fellowships will be awarded to selected candidates of British nationality with degree or equivalent qualifications to enable them to obtain wider experience and additional training in industrial establishments, at home or abroad, to make them more suitable for future employment in senior technical and administrative positions in British metallurgical industries. Each Fellowship will cover one full working year. Applicants will be required to state details of the programme they wish to carry out. Particulars and forms of application are available from The Secretary, Mond Nickel Fellowships Committee, 4 Grosvenor Gardens, London, S.W.1. Completed application forms are required by 1st June.

NEWS OF SCIENCE AND INDUSTRY

Symposium on Protection of Motor Vehicles

"The Protection of Motor Vehicles from Corrosion" is the subject of a symposium to be held at the Institution of Mechanical Engineers, 2 Birdcage Walk, London, S.W.1, on Tuesday and Wednesday, 11th and 12th March. Ten papers are being presented at the symposium, which is organized by the Corrosion Group of the Society of Chemical Industry. Preprints will be available for those who register as participants; the conference fee is 10*s*. for members of the Society and £2 for non-members.

Further particulars and registration forms may be obtained from the Society of Chemical Industry, 14 Belgrave Square, London, S.W.1.

Plaque to Charles Manby

A commemorative tablet is to be set up at 10 Lower Grosvenor Place, London, S.W.1, to Charles Manby, F.R.S. (1804–1884). A Committee of the Newcomen Society is administering the project, which has received financial support from The Iron and Steel Institute.

Manby was entrusted with the building and trials of the world's first iron steamer, the P.S. *Aaron Manby*; the parts of this vessel were built in 1821 at the Horseley Ironworks, Tipton, Birmingham, in an establishment conducted by his father, after whom it was named. In the following year he joined the Charenton works of Manby and Co., near Paris; while in France, he undertook the reorganization of the Le Creusot steelworks.

Charles Manby returned to the U.K. in 1830, to take over the management of the Beaufort Ironworks in South Wales; he was later associated with the Ebbw Vale Iron Co., where he introduced some modifications in the rolling of rails. He was Secretary of the Commission which in 1855 approved the Suez Canal Project. He was elected a Fellow of the Royal Society in 1853.

Metallurgy and Ski-ing

The 4th Metallurgical Ski-ing Meeting is being held at St. Christoph/Arlberg, Austria, from 8th to 13th April. Details may be obtained from the Department of Metallurgy, Montanistische Hochschule, Leoben, Austria.

Metallurgy in Glasgow

The Journal of the Metallurgical Club of the Royal College of Science and Technology, Glasgow, is being published at the end of March. It will contain articles on the kinetics of slag–metal reactions, the sampling of steel for oxygen determination, and the use of the electron microscope in crack studies. Copies are available, price 5*s*. each, from the College, Glasgow, C.1.

Retirement of V.d.Eh. Librarian

Dr. H. Dickmann, Librarian of the Verein deutscher Eisenhüttenleute, retired from that position at the end of 1957. He will continue to work with the Verein on matters concerning the history of the iron and steel industry.

Corrigenda

In the paper by the B.I.S.R.A. Methods of Analysis Committee entitled "The Combined Determination of Niobium and Tantalum in Steel" (*J.I.S.I.*, 1957, December, vol. 187, pp. 341–343), the chromium content given in Table I for sample 234 should be 18·0%.

DIARY

3rd Mar.—Cleveland Scientific and Technical Institution—"Iron and Steel Works Maintenance," by A. V. Williams—Cleveland Scientific and Technical Institution, Middlesbrough, 6.30 P.M.

3rd Mar.—Institute of British Foundrymen (Sheffield and District Branch)—"Foundry Sands," by W. B. Parkes—College of Commerce and Technology, Pond Street, Sheffield, 7 P.M.

4th Mar.—Institute of Metals (Oxford Local Section)—"Nucleation and the Cast Structure," by Dr. V. Kondic—Cadena Café, Cornmarket Street, Oxford, 7 P.M.

4th Mar.—Institute of Metals (South Wales Local Section)—"Metals for High Temperature Service," by Dr. W. Betteridge—Department of Metallurgy, University College, Swansea, 6.30 P.M.

4th Mar.—Sheffield Metallurgical Association—"The Application of Stable and Radioactive Isotopes in Analytical Chemistry," by A. A. Smales—B.I.S.R.A., Hoyle Street, Sheffield, 7 P.M.

6th Mar.—Institute of Metals (Birmingham Local Section)—"The Metallurgy of Steel for Deep Drawing and Pressing—Part II," by A. J. K. Honeyman—College of Technology, Gosta Green, Birmingham, 6.30 P.M.

6th Mar.—Institute of Metals (London Local Section)—"High Purity Iron," by B. E. Hopkins—17 Belgrave Square, London, S.W.1, 6.30 P.M.

6th Mar.—Leeds Metallurgical Society—"Some Metallurgical Problems of Nuclear Energy," by Dr. H. K. Hardy—Chemistry Wing, The University, Leeds, 7.15 P.M.

8th Mar.—Swansea and District Metallurgical Society—"Installation and Maintenance of Open Hearth Furnace Bottoms," by S. R. Isaac—Central Library, Swansea, 6.30 P.M.

10th Mar.—East Midlands Metallurgical Society—"Continuous Casting of Steel," by G. Fenton—College of Art, Green Lane, Derby, 7.30 P.M.

10th Mar.—Institute of Metals (Scottish Local Section)—"A Metallurgist in the Chemical Industry," by L. Powell, followed by Annual General Meeting—39 Elmbank Crescent, Glasgow, C.2, 6.30 P.M.

10th Mar.—Lincolnshire Iron and Steel Institute—Annual General Meeting—North Lindsey Technical College, 7 P.M.

10th Mar.—North East Metallurgical Society—"Recent Advances in Metallography," by Dr. D. McLean—Cleveland Scientific and Technical Institution, Middlesbrough, 7.15 P.M.

11th Mar.—Sheffield Metallurgical Association—Biennial Dinner.

11th and 12th Mar.—Society of Chemical Industry (Corrosion Group)—Symposium on "Protection of Motor Vehicles from Corrosion"—Institution of Mechanical Engineers, Birdcage Walk, London, S.W.1.

12th Mar.—Manchester Metallurgical Society—"Martensitic Transformations," by Dr. J. W. Christian—Central Library, Manchester, 6.30 P.M.

12th Mar.—North Wales Metallurgical Society—"The Work Hardening of Metals," by Professor Hugh O'Neill—Flintshire Technical College, Connah's Quay, near Chester, 7.15 P.M.

14th Mar.—Newport and District Metallurgical Society—"Plastic Coated Steel," by S. S. Carlisle—Whitehead Institute, Cardiff Road, Newport, 7 P.M.

14th Mar.—West of Scotland Iron and Steel Institute—"Use and Application of the Quantovac," by S. Muir—39 Elmbank Crescent, Glasgow, 6.45 P.M.

17th Mar.—Sheffield Society of Engineers and Metallurgists—"The Engineering Aspects of the Design of Nuclear Reactors," by E. P. Hotchen—Engineering Lecture Theatre, University Buildings, St. George's Square, Sheffield, 7.30 P.M.

18th Mar.—British Nuclear Energy Conference—"Further Developments in Nuclear Power," by Sir John Cockcroft (*James Forrest Lecture*)—Institution of Civil Engineers, 1–7 Great George Street, London, S.W.1, 5.30 P.M.

18th Mar.—Institute of Metals (South Wales Local Section)—Annual General Meeting—Department of Metallurgy, University College, Swansea, 6.30 P.M.

18th Mar.—Sheffield Metallurgical Association—"Trends in Basic Refractories," by H. Parnham—B.I.S.R.A., Hoyle Street, Sheffield, 7 P.M.

19th Mar.—Powder Metallurgy Joint Group (The Iron and Steel Institute and The Institute of Metals)—Informal discussion on "Developments in the Practice of Compacting and Sintering"—Church House, Westminster, London, S.W.1.

20th Mar.—Ebbw Vale Metallurgical Society—"The Impact of Nuclear Power on the Selection and Application of Material," by W. E. Dennis (Joint meeting with The Iron and Steel Institute)—Richard Thomas and Baldwins Lecture Room, 7.15 P.M.

20th Mar.—Staffordshire Iron and Steel Institute—"Steel: Its Finances and Economics," by H. E. Wincott—Station Hotel, Dudley, 7.30 P.M.

21st Mar.—Ebbw Vale Metallurgical Society—Annual Dinner.

22nd Mar.—Swansea and District Metallurgical Society—Annual Dinner.

24th–27th Mar.—Physical Society—Exhibition of Scientific Instruments and Apparatus—Caxton Hall, London, S.W.1.

25th Mar.—Sheffield Metallurgical Association—"The Acid Bessemer Process of Today," by F. B. Cawley—B.I.S.R.A., Hoyle Street, Sheffield, 7 P.M.

26th Mar.—Institute of Fuel (Yorkshire Section)—Annual General Meeting—Royal Victoria Hotel, Sheffield, 2.30 P.M.

28th Mar.—Society of Chemical Industry (Corrosion Group)—"Recent Research on the Corrosion of Boiler Tubes," by E. C. Potter—Robinson Lecture Theatre, The University, Manchester, 6.30 P.M.

CO-OPERATIVE TRANSLATION SERVICE

The following translations are now available, in addition to those given on p. 168 of the February, 1958, issue of the *Journal*:

533. Kazantsev and Strelets: "Temperature Distribution in an Ingot during Heating in a Soaking Pit." *Stal'*, 1957, No. 4, pp. 358–361. (£2 0*s*. 0*d*.)

534. Plankina: "Durability of Ingot Moulds at the Kuznetsk Metallurgical Combine." *Stal'*, 1957, No. 4, pp. 362–365. (£2 15*s*. 0*d*.)

618. Bungardt and Preisendanz: "Vapour Pressure Experiments on Fe–Cr Alloys." *Arch. Eisenhüttenwesen*, 1956, vol. 27, Nov., pp. 715–724. (£5 0*s*. 0*d*.)

627. Samsonov and Latysheva: "Diffusion of Boron and Carbon in Some Transition Metals." *Fizika Metallov i Metallovedenie*, 1956, vol. 2, pp. 309–319. (*Free*).

652. Heine: "The Present Position of Cold-heading, Cold-pressing and Flow-pressing in Bolt Manufacture." *Werkstattstechnik und Maschinenbau*, 1950, vol. 40, No. 6, pp. 214–220. (£4 0*s*. 0*d*.)

687. Tanoue *et al.*: "Research on the Hydrogen Analysis of Iron and Steel by the Low-pressure and Constant-volume Method." *Sumitomo Metals*, 1956, vol. 8, Oct., pp. 55–60. (£3 0*s*. 0*d*.)

701. Tsygankov: "Effect of the Shape of the Edges of Strip on the Welded Seam Strength of Tubes." *Stal'*, 1957, Aug., pp. 728–730. (£1 15*s*. 0*d*.)

704. Reininger: "Testing and Evaluation of Binders for Foundry Sands." *Giesserei-Praxis*, 1957, vol. 75, Apr. 25th, pp. 176–181. (£2 15*s*. 0*d*.)

726. Granovsky: "The Rolling of Spheres (Ball Bearings and Crusher Balls)." *Stal'*, 1956, No. 4, pp. 333–337. (£1 15*s*. 0*d*.)

728. Poncelet: "Study of Various Continuous Billet Mills." *Centre Doc. Sidér., Circ. Inform. Tech.*, 1956, No. 12, pp. 2425–2442. (£3 5*s*. 0*d*.)

730. Slutskaya: "13KH3GNMA Welding Wire for the Automatic Welding of 30KLGSA Steel." *Avtomaticheskaya Svarka*, 1956, No. 2, Mar./Apr., pp. 12–17. (£1 15*s*. 0*d*.)

736. Lifshits and Sochan: "Oxygen Utilization in a 250-ton Open-hearth Furnace." *Stal'*, 1957, May, pp. 402–405. (£2 0*s*. 0*d*.)

740. Matveev: "The Tube Industry of the U.S.S.R." *Stal'*, 1957, Nov., pp. 997–1005. (£4 10*s*. 0*d*.)

746. Papier: "Contribution to the Study of the Electrolytic Extraction of Carbide from Free-cutting Steel." *Rev. Mét.*, 1954, Oct., pp. 723–734. (£5 10*s*. 0*d*.)

750. Bakhtinov and Shternov: "Ways for Further Development of Reduction and Section Rolling Mills." From "Problems of Metallurgy," pp. 365–377: 1953, Moscow, U.S.S.R. Acad. Sci. (£4 10*s*. 0*d*.)

751. Boulange and Biau: "The Continuous Wire Rolling Train at the Wendel Plant at Jouef." *La Technique Moderne*, 1957, July, pp. 63–65. (£1 10*s*. 0*d*.)

752. Boulez: "The Continuous Wire Mill of the Société Métallurgique de Normandie." *La Technique Moderne*, 1957, July, pp. 66–69. (£1 10*s*. 0*d*.)

764. Mil'ner and Feodos'eva: "A Magnetic Method for Detecting Diffusion in Metals." *Fizika Metallov i Metallovedenie*, 1956, vol. 3, No. 3, pp. 483–485. (£1 5*s*. 0*d*.)

765. Gulyaev: "Study of Non-metallic Inclusions in Castings by the Use of Radioisotopes." *Liteinoe Proizvodstvo*, 1956, Aug., pp. 25–27. (£2 0*s*. 0*d*.)

772. Wever *et al.*: "Studies of Converter Smoke with a view to Spectroscopic Control of the Basic Bessemer Process." *Stahl u. Eisen*, 1957, vol. 77, No. 21, Oct. 17th, pp. 1451–1459. (£4 15*s*. 0*d*.)

773. Koch and Malissa: "Sensitive Determination of Carbon with a Recording Conductivity Instrument." *Arch. Eisenhüttenwesen*, 1956, vol. 27, Nov., pp. 695–700. (£2 15*s*. 0*d*.)

777. Drahokoupil: "Rapid Determination of Tungsten in Steels by means of Secondary X-ray Emission Analysis." Techna Physical Institute, CSAV, Prague. (£3 0*s*. 0*d*.)

793. Dazai: "The Modern 4-strand High-speed Continuous Wire Rod Mill at Hikari Works." *Tetsu to Hagane*, 1956, vol. 42, Sept., pp. 935–936. (£1 15*s*. 0*d*.)

809. Castro *et al.*: "A Source of Error in the Chemical Determination of Nitrogen in Special Steels." *Analytica Chimica Acta*, 1957, vol. 17, No. 6, Dec., pp. 530–534. (£1 0*s*. 0*d*.)

833. Nekrassov: "Soviet Blast-furnace Production", *Stal'*, 1957, Nov., pp. 965–968. (£2 10*s*. 0*d*.)

Note: When two copies are required, there will be 50% reduction for the second copy. Six copies can be bought for the price of two, and twelve copies for the price of three.

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects

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MINERAL RESOURCES

Nature and Origin of Southwestern Oregon Chromite Deposits. L. Ramp. (*Min. Eng.* 1957, **9**, Aug., 894–897).

Geology and Mineral Resources of the Congonhas District Minas Gerais, Brazil. P. W. Guild. (*U.S. Geol. Survey Professional Paper* 290, 1957, pp. 90). An account of iron mining districts, reserves and mines and a survey of ores is included, also manganese.

Developments in Geophysics and Geochemistry. Application to the Search for Metallic Deposits. L. Neltner and P. Lasfargues. (*Rev. Ind. Min.*, 1957, Sept., 773–807). The authors discuss the principles, apparatus and results for various methods of prospecting for metallic deposits. The techniques include gravimetric magnetic methods, also aerial survey and the nuclear magnetometer. The limitations of each method are discussed.—R. P.

The Lognormal Theory of the Systematic Sampling of Mineral Deposits. G. Matheron. (*Ann. Mines*, 1957, Sept., 566–584). The theory previously advanced is extended to cases where the sampling is not strictly random.

ORES—MINING AND TREATMENT

Requirements for Ships for the Export of Ores from Overseas Territories. R. Pré. (*Ann. Mines*, 1957, Sept., 606–610). Tables of ores and minerals of all kinds, including iron ores, are estimated for the years 1957–1963 from French overseas territories to various destinations.

Studies of Hydrated Iron Oxides. I. Hydrated Ferric Oxide Minerals Altered from Pyrites by Weathering. M. Nambu. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1957, **9**, June, 215–226). Composition of limonites was investigated. Goethite and lepidocrocite were found with occasionally hydrated haematite and amorphous hydrated Fe_2O_3.

Oxidation of Iron Oxides. F. Lecznar. (*Hutnik*, 1956, **23**, Nov., 413–418). [In Polish]. The physicochemical properties of FeO, Fe_3O_4 and γ-Fe_2O_3 are discussed. Experimental investigation of the oxidation of limonites contaminated with siliceous matter is described. The siliceous limonite reduced to magnetite by heating at 750° C, 850° C, and 950° C and cooled either in a furnace or in air oxidized to maghemite. Limonite reduced to FeO at 850° C and 950° C and cooled in air, oxidized to magnetite. Fe_3O_4 and FeO cooled in the range 850–950° C showed the smallest losses of iron in scrap.

On the Mechanism of Oxidation of Magnetite. V. T. Arkharov and B. S. Borisov. (*Doklady Akad. Nauk S.S.S.R.*, 1957, **114**, (2), 293–296). [In Russian]. Monocrystals of magnetite oxidized at temperatures above 800° C were studied by X-ray and electronographic methods. The fine structure and the crystalline phases of the products of oxidation are discussed in detail.—S. I. T.

On the Structure of Magnetite Within the Reaction Diffusion Layers During the Reduction of Haematite. V. I. Arkharov and V. N. Bogoslovskii. (*Fizika Metallov i Metallovedenie*, 1956, **(2)**, 254–268). [In Russian]. It has been shown experimentally that in the first stage in the reduction of haematite, there appears the triclinic lattice of the spinel phase owing to the distortion of the cubic lattice, but that it gradually alters as the reduction proceeds to the γ-phase, i.e. to Fe_3O_4. The experiments have supplied the highest probability distribution of ions of various valencies in the lattice spaces of the γ-phase and have also indicated the probable mechanism of the transformation of the haematite into magnetite. Finally they have pointed the way to the possibility of regulating the kinetic transformation of iron oxides and the heat stability of iron alloys.—L. H.

Example of the Application of Soil Pressure Measurement in the Iron Mines of Lorraine. Tincelin and Sinou. (*Rev. Ind. Min.*, 1957, Sept., 820–844). After reviewing several methods of exploitation, the authors describe work carried out to study suitable supports and prevention of subsidence, using sections of the mine. The factors governing the erection of these supports are discussed, also the mechanical strength of the rocks and soil above and around the sections. Probable pressure distribution is considered.—R. P.

Dredging at Steep Rock Lake. P. D. Pearson. (*Min. Eng.*, 1957, **9**, Aug., 898–902). An illustrated account of overburden removal from the iron ore body is given.

Structural Conditions at the Wabana Iron Mines, Newfoundland. D. K. Norris. (*Canad. Min. Met. Bull.*, 1957, **50**, Sept., 539–541).

Mining and Usage of Molybdenum. Climax Molybdenum Co. (*Min. J.*, 1957, **249**, Sept. 6, 274). A brief account of mining methods at Climax, Colorado and notes on the uses of Mo are given.

Study of Reducibility and Strength of High Siliceous Sintered Iron Ore Lumps. T. Ikeno. (*Tetsu to Hagane*, 1957, **43**, Mar., 213–214). [In Japanese]. Variations of reducibility, strength and FeO content of sinter with coke content

are reported for various ores, principally of the goethite type.—K. E. J.

Slotted Screens for Ore Beneficiation. I. N. Nedovizii. (*Stal'*, 1957, (9), 850–854).

Dry Magnetic Concentration. P. E. Cavanagh and E. W. Williams. (*Canad. Min. Met. Bull.*, 1957, **50**, *Sept.*, 558–564). Treatment of low-grade magnetite with the Aerofall mill and other equipment is explained.

Beneficiation of a Lateritic Iron Ore from Rajhara Pahar, Madhya Pradesh. S. B. Das Gupta, G. V. Subramanya and P. I. A. Narayanan. (*J. Sci. Indust. Res.*, 1957, **16A**, Aug., 373). A letter giving the mineralogical make-up of the ore and the effects of unsuccessful attempts to produce concentrates.

A Study on the Utilisation of Laterite. On the Recovery of the Nickel in Laterite. M. Tanaka. (*Tetsu to Hagane*, 1957, **43**, Mar., 269–270). [In Japanese]. Experiments were made on ores containing 0·66% and 0·98% Ni. They were H_2-reduced, forming Ni and iron, then leached with NH_3–$(NH_4)_2$ CO_3 solutions under oxidizing conditions, forming $NiCO_3m$ $3NH_3$ and ferric hydroxide. Extraction figures up to 80% of the Ni are reported.—K. E. J.

Development of the Hydrocyclone. S. E. Erickson. (*Min. Eng.*, 1957, **9**, Aug., 869–872). Application to mineral beneficiation is discussed. **Selecting a Cyclone for Wet Classification.** E. C. Herkenhoff, (873–876). Types and capacities and particularly factors and purposes governing choice are given. **Compilation of Cyclone Operating Data.** T. M. Morris. (877–879). **Use of Cyclones in the Grinding of Taconite.** F. D. De Vaney. (880–882). Practice in Minnesota is outlined. **Cyclone Practice in Arizona.** R. Salter and E. J. King. (883–889).

Something New in Sintering. H. E. Rowen, P. V. Gallagher, and T. E. Ban. (*Iron Steel Eng.*, 1957, **34**, Aug., 133–144). The authors review developments in sintering techniques from the early days of sintering to the latest developments brought about in the U.S.A. and Europe. Technical data and general arrangement drawings are given.—M. D. J. B.

Instrumentation for a Sinter Plant. A. A. Latowski. (*Indust. Heating*, 1957), **24**, Aug., 1584, 1586, 1688). Control of blending, moisture, ignition and combustion is outlined.

Design of Rational Ventilating Arrangements for Sinter Plants. S. S. Khalkin. (*Stal'*, 1957, (3), 205–208). [In Russian]. After a general discussion of working conditions in iron-ore sintering plants, the results of air dust- and CO_2-content and temperature determinations at two Soviet plants are presented together with other plant data. Measures taken and recommended for improving conditions are considered.—S. K.

Determination of the Retention Time of Materials in a Ball Iron Rotary Kiln by the Radioactive Tracer Method. V. F. Knyazev and O. V. Travin. (*Zavodskaya Laboratoriya*, 1956, **22**, (9), 1071–1072). [In Russian]. An account is given of transit-time determinations in a semi-industrial scale rotary kiln fired with coke-oven gas at the Orsk-Khalilovskii Metallurgical Combine. ^{45}Ca in the form of oxide was introduced in a glass container with the charge, and activity determinations were made on samples taken after the lapse of 4 h. Results of four experiments are shown graphically and discussed in relation to productivity and state of the kiln. The kiln is rated at 110–125 kg. ore/h, its length, diameter and inclination are 8 m, 1·02 m, and 2%, respectively, and it performs 0·5 r.p.m.—S. K.

Ferro-Coke and Its Production. F. Roll. (*Met. Constr. Masini*, 1956, **8**, (3), 41–44). For an economic yield from poor iron ore, cheap fuel is essential. Hence the work initiated at the Thyssen Gas and Water works in Duisburg and the tests carried out on an industrial scale in the Soviet Union are considered in great detail, and recommendations are made for their application and adaptation to the conditions prevailing in Rumania.—L. H.

Studies on Electric Pig-Iron Smelting of Iron Sand. I. Experimental Studies of Lime Sintering. K. Takai. (*Tetsu to Hagane*, 1957, **43**, Mar., 372–373). [In Japanese]. For mixtures containing varying proportions of lime, the min. charge density is found with approx. 7% of moisture, and the max. permeability with 10–14%. A linear relationship is found between permeability and sintering velocity. The microstructure of sinters with added lime is studied from photomicrographs.—K. E. J.

Sintering Test with Open-Hearth Slag Addition. A. Chida. (*Tetsu to Hagane*, 1957, **43**, Mar., 354–355). [In Japanese]. Details are given of the sinter mixes and the size and chemical analyses of the slag additions. Additions up to 15% are shown to increase the sinter strength but to lower reducibility.

Improved Method of Determining Softening Temperatures of Ores and Sinters. A. G. Astakhov. (*Zavodskaya Laboratoriya*, 1956, **22**, (5), 571–572). [In Russian]. In the apparatus described the movement of a loaded rod as it sinks into the heated sample of ore or sinter is converted into electrical changes. These are recorded on the same diagram as the temperature.—S. K.

FUEL—PREPARATION, PROPERTIES, AND USES

Progress in the Understanding of Combustion. A. D. Cummings. (*J. Inst. Fuel*, 1957, **30**, Sept., 499–505). Historical.

The Application of Oxygen Analysis to Combustion Control. J. C. Farquhar. (*Instrument Eng.*, 1957, **2**, Oct., 65–74). The use of magnetic continuous analysers in control of combustion processes is evaluated. Flue-gas analysis is considered and fuel : air and steam : air ratio controls are compared.

A Note on Coal Sampling Theories. R. C. Tomlinson. (*Fuel*, 1957, **36**, Oct., 442–446). Variance of sampling for ash content is formulated and the Hassialis formula is shown to be equivalent to the simpler expression of Kassel and Guy. Conditions of random sampling and mixing are rarely met with in practice.

Determination of Froth-Flotation-Washability Data. C. C. Dell. (*J. Inst. Fuel*, 1957, **30**, Sept., 523–526). A mathematical treatment for minerals is extended to coal.

Main Directions of the Utilization of Coals of the Kuznetsk Basin in Relation to the Origin of their Organic Mass and Mineral Admixtures. M. Yu. Grigor'ev and G. N. Podbel'skii (*Izvest Akad. Nauk. S.S.S.R. Otdelenie Tekh. Nauk*, 1956, (10), 98–107). [In Russian]. A classification diagram of the washability of coals of the Kuznetsk basin is proposed. The coals are classified on the basis of the yield and ash content of the concentrate and the yield of products of dens. 1·5–1·8, depending on the origin of mineral inclusions and the conditions of formation of coal strata.—V. G.

An Application of Linear Programming to the Transport of Coking Coal. A. H. Land. (*J. Roy. Stat. Soc.*, 1957, **120A**, (3), 308–319). Coal available at 154 collieries and required at 65 coke ovens is considered and it is estimated that a saving of 10% in ton-miles could be attained by linear programming.

Preparation of the Charge for Coke Production. I. Barbu, S. Forizs, and D. Miortescu. (*Met. Constr. Masini*, 1957, **3**, 27–31). [In Rumanian]. The main point made is the influence of the size of coal on the production of coke. The crushing in Rumania is not uniform. In Reshitza, the coals are crushed once only giving between 80–85% under 3 mm. This work describing parallel experiment in the laboratory and in a pilot plant emphasizes this point with tabulated results and statistical data.—L. H.

Viscoelastic Properties of Coal During Carbonization. D. Fitzgerald. (*Fuel*, 1957, **36**, Oct., 389–394). Work with a Gieseler plastometer shows the plasticity of coal during carbonizing to be not purely viscous and a model is advanced. The values of the 4 parameters change and values for one coal are given during constant-temperature carbonization. Relation to structure is indicated.

Carbonized Fuels from Non-Coking Coals. D. W. Gillings. (*Year Book Coke Oven Man. Assoc.*, 1957, 175–209). Plastometry and blending tests with special reference to briquette carbonizing processes of " Furnacite " type are reviewed, and an account of fluidized carbonization and hot sand heating on the laboratory scale given.

Production of Coke from Unsuitable Coals. I. Barbu and I. Stefanescu. (*Met. Constr. Masini*, 1957, **9**, (1), 41–49). [In Rumanian]. After reviewing the various methods propounded for coking unsuitable coals the two stage and the stage-and-a-half processes used in Rumania are described in detail. Results are also given of cast iron produced from such coke in a cupola of small capacity.—L. H.

Briquetting of Indian Coals: Part IV. Carbonization of Lignite Briquettes. D. N. Sibal, R. S. Ghosh, and M. S. Iyengar. (*J. Sci. Ind. Res.*, 1957, **16A**, July, 321–323). Compressive

strength of lignite briquettes and effects of moisture and rate of heating are recorded.

Briquetting of Indian Coals: V—Briquetting Performance of Coals of Different Rank. M. S. Iyengar and T. A. Subramanian. (*J. Sci. Indust. Res.*, 1957, **16A**, Aug., 359–365). Variables in briquetting and their effects on high, medium, and low rank coals are investigated.

Microhardness of Carbonized Coal. H. Honda and Y. Sanada. (*Fuel*, 1957, **36**, Oct., 403–416). Measurements on 14 coals are reported after carbonizing at 200–2000° C. Changes with temperature, composition and rank are indicated. Vickers and Knoop hardnesses are considered and Young's moduli are deduced.

Improving the Quality of Foundry Coke. M. V. Goftman and V. I. Butorin. (*Liteinoe Proizvodstvo*, 1956, (11), 5–7). [In Russian]. After indicating the importance of foundry coke of good quality and the frequent failure of Soviet foundries to obtain this, the authors describe experiments with laboratory and semi-works scale equipment in which the effects of changes in coking charge composition on coke quality were studied. Good cokes were obtained when anthracite was added to the coking charge, especially when coking speed was increased. With anthracite contents of 25% the pitch requirements (10%) became excessive; pre-compression of the charge heated to the pitch softening temperature enabled pitch content to be kept down to, e.g., 7–10% with an anthracite content of 70–80%.—S. K.

The Collin Coke Oven. M. S. Gaskill. (*Year Book Coke Oven Man. Assoc.*, 1957, 367–382).

Characteristics of Turbulent Jet Diffusion Flames. M. W. Thring and E. H. Hubbard. (*Institute of Fuel Preprint*, 1957, Oct. 9, pp. 20). Studies of atomized oil and high pressure coke-oven gas flames in experimental furnaces are summarized and relationships for entrainment and luminosity are given and flame temperature distribution is discussed.

Heat Transfer from Flames in the Steel Industry. J. H. Chesters and R. Mayorcas. (*Institute of Fuel Reprint*, 1957, Oct. 9, pp. 26). Work to link model studies with full-scale O.H. furnace operation was carried out at Ijmuiden and the use that can be made of the investigations is outlined. Work on burners has led to understanding of atomization and jet momentum, the part played by C in conferring luminosity and the effects of C/H ratio have been studied and application of the results of calculations to full scale working are set out.

Bacteriological Treatment of Coke Oven Effluents. H. A. Page. (*Year Book Coke Oven Man. Assoc.*, 1957, 441–467). An account of the plant at Altham Coke Works and its development is given.

AIR POLLUTION AND SMOKE

Purification of Air in the Metallurgical Industry and the Selection of the Purification System. D. Strominger. (*Met. Constr. Masini*, 1957, **9**, (2), 74–80). [In Rumanian]. A well-known method of air purification is described, and tables are given to obtain at a glance the data determining the plant required. A graphic method is also given to help in the selection of the most suitable installation and also a calculation to obtain approximately the efficiency of purification under the given conditions.—L. H.

Small Size—Better Efficiency Feature New Dust Collector. (*Iron Steel Eng.*, 1957, **34**, July, 151–152). A wet inertial type dust collector is described which is claimed to have a greater efficiency than any comparable wet type collector.

The Prevention of Grit and Dust Emission. C. J. Stairmand. (*Smokeless Air*, 1957, Autumn, 51–60). A review of types of dust arrestors and their efficiencies is given.

The Cupola Air Pollution Problem. T. L. Harsell, jun. (*B.C. Professional Eng.*, 1957, **8**, July, 16–18, 20–22). Control measures in Los Angeles are described and operating conditions for 3 cupolas are tabulated with measurements and analyses of gases emitted and bases for the design of anti-pollution equipment are given. The legal background is sketched.

Air and Water Pollution. The Position in Europe and in the United States. (*O.E.E.C. Project* Nos. 136 et 145, 1957, Feb., pp. 218). Legislation is outlined and visits to plants, including mining and ore, coal and coke (but not metallurgical) plants are reported with reference to water pollution. Air pollution legislation is then reviewed and visits to metallurgical and other plants are surveyed, with notes on the effects of air pollution. Numbers of steel plants were inspected, but the accounts given are very brief.

TEMPERATURE MEASUREMENT AND CONTROL

Primary Elements for Temperature Measurement. H. G. Oughton. (*Instrument Eng.*, 1957, **2**, Oct., 76–82). Resistance, thermocouple and radiation pyrometers are reviewed.

Immersion Thermocouple Sheaths. O. M. Margulis, K. G. Romanchenko, and I. A. Getman. (*Stal'*, 1957, (8), 714–715). Ceramic sheaths were found to crack at the liquid metal-air interface. A new type for immersion for up to 4 h is described based on ZrO_2 stabilized with CaO and magnesite.

Results of the Calibration of Thermocouple Pyrometers used in a Steel Works. M. Jindai. (*Tetsu to Hagane*, 1957, **43**, Mar., 396–398). [In Japanese].—K. E. J.

Liquid Steel Temperature Measurement. W. C. Heselwood and D. Manterfield. (*Platinum Metals Review*, 1957, **1**, Oct., 110–118). A review of the quick-immersion thermocouple method is given with principles of design, use and development, maintenance and defects and sheath composition as headings.

Measurement and Recording of Temperatures during the Pouring and Solidification of Steel. L. Šmrha and R. Hluzin. (*Hutnické Listy*, 1957, **12**, (8), 711–714). Good results were obtained with Pt/PtRh couples in quartz capillary tubes and a mechanical high-speed recorder was developed which is superior to the loop-oscillograph unless the changes are very rapid. The recorder can be arranged for differential thermal analysis when its usefulness is extended further.

Organization of Liquid Steel Temperature Control with Thermocouples in the Steel Melting Shops of the Kuznetskii Metallurgical Combine. A. J. Vasil'ev, N. F. Novogorodtsev, S. G. Otlivanov and G. G. Tereshin. (*Zavodskaya Laboratoriya*, 1956, **22**, (9), 1127–1130). [In Russian]. At present 5000 steel-temperature determinations are made monthly at the Kuznetskii Metallurgical Combine by immersion of sheathed Pt–Pt-Rh couples. The construction of the hot end of the equipment is described and the accuracy of a large number of measurements is analysed. The relative frequencies of defective measurements due to different causes are tabulated. The staff involved in the preparation, use and maintenance of the equipment consist of a foreman, a senior electrician and 15 shift electricians. The cost per measurement is given as 10 roubles, the cost per ton of steel being 0·20 roubles.—S. K.

Method for the Measurement of the Temperature of Bodies Rotating at High Speeds. B. A. Furs. (*Liteinoe Proizvodstvo*, 1956, (12), 19–20). [In Russian]. The method described has been used for measuring the temperature of rolls and of the mould in centrifugal casting. Thermocouples are used, connected to the instrument through a ring-and-brush arrangement.—S. K.

REFRACTORY MATERIALS

The Possibility of Predicting Theoretically The Resistance to Heat of Refractory Materials. L. Massimilla and G. Astarita. (*Calore*, 1957, **28**, June, 244–247). [In Italian]. This article summarizes the results obtained over several years from studies carried out at the Institute of Industrial Chemistry of The University of Naples on the resistance of refractories submitted to stresses of thermal origin. (11 references).—M. D. J. B.

Symposium on High Temperature Refractories. K. D. Bartels, E. E. Callinan, and E. R. Sullivan. (*Iron Steel Eng.*, 1957, **34**, July, 106–117). The symposium consisting of 3 papers covers the subject of refractory bricks in O.H. furnaces, various types of reheating furnaces and in soaking pits.

Manufacture of Refractory Cements in Rumania. N. Deica. (*Met. Constr. Masini*, 1956, **8** (2), 48–57). Full explanations are given for the production of four groups of refractory cements suitable for use at temperatures between 900°–1550° C. The admixtures with the cements consist of such substances as magnesite, bauxite, alumina, chromite etc. which are found in the country in substantial quantities.

A Contribution to the Study of Forsterite Refractory Materials. V. Siniansky and L. Solomon. (*Met. Constr. Masini*, 1956, **8**, (12), 30–34). [In Rumanian]. This is a full description of the technique of the manufacture in the laboratory

and on a semi-industrial scale. No practical conclusions are drawn.—L. H.

The Application of Alumino-Silicate Refractories in the Foundry Industry in the United Kingdom. J. H. Myrtle. (*Inst. Brit. Found.*, 1956, **7**, 15–34). Refractories, volume stability, thermal shock resistance, resistance to slags and fluxes and high-temperature strength are briefly discussed. Types of brick, castable and mouldable refractories, foundry uses for steel and cast iron, in cupolas and electric furnaces and for non-ferrous metals are reviewed. Notes on pyrometry and foundry accessories are added.

High Temperature Properties of Mould Refractories. J. White and J. L. Evans. (*Castings*, 1957, **3**, June, 15, 17). Effects of heat on clay-bonded sands and on Wyoming bentonite are studied.

Muffle Furnace for Testing the Resistance of Refractory Products to Heat Shock. T. I. Dumitrescu. (*Met. Constr. Masini*, 1956, **8**, (3), 48–49). A gas heated furnace has been selected for economy and convenience. Materials for electric furnaces have to be imported whereas methane gas is available in large quantity, hence the advantage of constructing the muffle. The furnace allows heating to 950° C within 40 min and then cooling down to 20° within 3 min. Its life is said to be between 1 and 2 years. Satisfactory results have been obtained and in consequence recommendations to the various authorities are made.—L. H.

Thermal Conductivity: XIII, Effect of Microstructure on Conductivity of Single-Phase Ceramics. F. R. Charvat and W. D. Kingery. (*J. Amer. Ceram. Soc.*, 1957, **40**, Sept., 306–315). Conductivities of Al_2O_3, MgO, TiO_2, and CaF_2 are measured between 0–1000° C. Effects of impurities and pores are noted.

Phase Relationships in the System Zirconia-Thoria. P. Dewey and E. Loh. (*J Amer. Ceram. Soc.*, 1957, **40**, Sept., 321–324).

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Processes for the Reduction of Iron Ores. G. Grenier. (*Echo Mines*, 1957, Aug., 458–460). A review naming the physico-chemical and mechanical principles and surveying very briefly the blast furnace with coke or charcoal, the low-shaft furnace, the electric furnace and Stürzelberg and Basset methods. Methods not employing fusion, Catalan, Sieurin, Wiberg-Söderfors and Renn-Krupp, are then outlined.

New Blast Furnaces on the North-East Coast. (*Iron Coal Trades Rev.*, 1957, **175**, Sept. 27, 721–725). An illustrated account of the new Dorman Long, Clay Lane Plant.

Rebuilding Blast Furnace No. 2. N. V. Krepyshev. (*Metallurg*, 1956, (2), 5). The furnace, at Kuznetsk, was increased in volume from 1163 to 1310 m³ and equipped with air cooling below the hearth using 4–6 in. steel pipes and consuming 16000–18000 m³/hr of air. An electric tap-hole gun with piston pressure of 160–200 t can close the iron notch while on full blast and at full top pressure.

A Typical 2286 m³ Blast Furnace. Its Mechanization and Automation. A. E. Sukhorukov. (*Metallurg*, 1957, (5), 5–8). A new furnace planned at Gipromez will operate on 100% self-fluxing sinter each lump size being charged separately. A 25% O_2 blast with steam will be used at a temperature of 1200° C and 1·8 atm top pressure. Output planned is 5000 t/day and the coefficient (working volume ÷ tonnes per 24 hr) will be as low as 0·46. Automatic control of charging, heating and humidifying the blast, removing iron and slag and feeding of all auxiliary materials to the furnace area will be applied. These are described in detail.—R. S.

The Loading of Blast Furnace Charges. Z. Krotkiewski. (*Hutnik*, 1957, **24**, Jan., 1–10). [In Polish]. The author gives a survey of methods of loading blast furnace charges so far used. A detailed description with diagrams of the experimental methods of T. A. Teschia is given. The author's own loading apparatus, which was patented in 1946 is fully described and a comparison of the two given.

Study on the Distribution of Burden Materials in the Blast Furnace. I. K. Kanbara. (*Tetsu to Hagane*, 1957, **43**, Mar., 270–271). [In Japanese]. Laboratory model experiments on the effects of various charging systems are described.

Development of Controlled Air Distribution for the Blast Furnace. J. M. Stapleton. (*Blast Furn. Steel Plant*, 1957, **45**, Sept., 1007–1017). Wear of the lining by uneven blast distribution is observed, though this is bound up with the charging of the stock. Work at Gary is reported in which temperature measurements were made and velocities approximately computed showing that peripheral velocity was far from uniform. Typical distributions are shown and a proportioning system was designed and its operation is indicated. Wear of furnace walls is shown and economies are indicated.

The Influence of Moisture in the Blast on the Process of Combustion of Coke in a Blast Furnace. A. A. Markin and M. Ya. Ostroukhov. (*Izvest. Akad. Nauk, Otdelenie Tekh. Nauk*, 1956, (11), 58–63). [In Russian]. An investigation of the influence of moisture in the blast (up to 50 g/m³) on the size of the combustion zone in front of tuyeres was carried out on a furnace of 700 m³ working volume producing foundry iron. The method consisted of sampling gases in front of tuyeres at various distances from the tuyere nozzle and their analysis. It was established that on increasing the amount of moisture by 10–20 g/m³ no change in the length of the combustion zone takes place. If the consumption of heat for the decomposition of the moisture added is not compensated by an appropriate increase in the blast temperature, the velocity of the combustion of coke decreases. The beneficial influence of moisture in the blast, (when it is not totally compensated by a rise in blast temperature) is explained by a decrease in the temperature on the boundary of the oxidizing zone. This decrease in the temperature is accompanied by a decrease in the volume of the gas i.e. its velocity.—V. G.

Economies in the Consumption of Coke Due to Improvements in the Construction and in the Working of Blast Furnaces with Three Tiers of Tuyeres. C. Wanyorek. (*Met. Constr. Masini*, 1956, **8**, (5), 39–45). The principle of this adaptation consists in introducing above the usual tuyeres a supplementary forced air current so as completely to reduce the CO resulting from CO_2 reduction (endothermic). This is achieved by building above the usual tuyeres two further rows of air vents. Claims are made that the economies in consumption of coke range from 20%–44%.—L. H.

Comparative Trials for the Utilization of Ankerite in Blast Furnaces. N. Pilli. (*Met. Constr. Masini*, 1957, **9**, (2), 47–56). [In Rumanian]. The increasing need for metal makes the utilization of ankerites essential; these have to be added to the normal charge for grey cast iron. Comparative tests have been made and data obtained which allow the drawing up of a balance sheet for the costs of raw materials and heat. The results confirm that roasted ankerite can be used to advantage in blast furnaces for the production of cast iron.—L. H.

Method of Checking the Specific Coke Consumption and Cast Iron Consumption in Blast Furnaces. V. Codreanu. (*Met. Constr. Masini*, 1956, **8**, (12), 57–62). [In Rumanian]. The problem proposed is to calculate the ratio of coke and cast iron in relation to the completely finished and machined castings under optimum conditions. From the theoretical point of view, formulæ have been calculated for the use in the respective productive units.—L. H.

Operation of Blast Furnaces on High Top Pressure. I. A. Kuz'min. (*Metallurg*, 1957, (7), 3–5). Two furnaces at the Cherepovets plant of 1007 and 1033 m³ capacity were designed for 1·4 atm top pressure. One has operated since April, 1956 at 1 atm. Use of self-fluxing sinter of basicity 1·15 and higher top pressure, improved efficiency. Coke consumption was reduced from 15 to 13 cwt/t iron and pressure is now being increased to 1·5 atm with a blast temp. of 950° C. Details of operation are given.

Operation of a Blast Furnace at up to 1·1 atm. Gauge Top Gas Pressure. V. P. Onoprienko, B. N. Starshinov, A. A. Tkachenko, V. D. Sinitskii, L. M. Freidin, and L. Ya. Portnyi. (*Stal'*, 1957, (9), 772–778). Raising top pressure from 0·6–0·7 atm to 1–1·1 atm gave improved performance. Taphole and slag notch plugging without decrease of blast or top pressure is possible.—R. S.

Experimental Use of Ferro-coke in the Blast Furnace. T. Tsuru. (*Tetsu to Hagane*, 1957, **43**, Mar., 351–352). [In Japanese]. Chemical and physical properties are given for coke and ferro-cokes made with 5% and 10% of ore fines. Comparative furnace operating results and iron analyses are given for a trial period of 14 days.—K. E. J.

A Model Experiment on Hanging in the Blast Furnace Caused by Flooding. I. K. Segawa. (*Tetsu to Hagane*, 1957, **43**, Mar., 352–354). [In Japanese].—K. E. J.

On the Relation between the Size and Form of Centre Core and the Furnace Conditions of the Blast Furnace. S. Ota. (*Tetsu to Hagane*, 1957, **43**, Mar., 216–218). [In Japanese]. Measurements of the distance between the edge of the central core and the furnace lining at the level of the slag notches enable the core shape and size to be estimated under various conditions of blast temperature, CaO/SiO_2 ratio, slag temperature and quality of iron made.—K. E. J.

Investigations of the Life of Carbon Linings in Blast Furnace Stacks. F. A. Khil'kevich and S. V. Bazilevich. (*Stal'*, 1957, (9), 769–771). A six months' trial showed C to be more suitable than fireclay and the thickness of the lining could be reduced to 20 in. (The editor comments that the trial was too short for this conclusion). External losses were not increased. In presence of Zn a double lining was necessary of C and vermiculite with temperatures of 1200° on the inside of the C, 600° at C-vermiculite junction and 10° at the cooler.—B. S.

On the Use of Carbon Blocks for Blast Furnaces. Y. Shiraishi. (*Tetsu to Hagane*, 1957, **43**, Mar., 264–265). [In Japanese]. The wear of the hearth blocks at Kukioka No. 3 furnace after 921 days' operation is shown, in relation to the tuyeres, slag notch and tap-hole. Changes in physical and chemical properties of the carbon after 921 days, as compared with those before blow-in, are tabulated. Diagrams are given of proposed new carbon hearths for Kukioka No. 2 and 3 furnaces (hearth diameters 7000 and 7700 mm, respectively).—K. E. J.

Comparative Study of Carbon Bricks in the Blast Furnace. E. Perez Blanco. (*Inst. Nac. Carbon Bol. Inf.*, 1957, **6**, Jan.-Feb., 16–21). Two types are studied in the blast-furnace hearth. Microscopical examination is reported from which the starting materials for each are inferred.

Study of the Blast Furnace Process by Means of Isotopes. J. P. Bardin, P. L. Gruzin and S. V. Zemskii. (*Doklady Akad. Nauk S.S.S.R.*, 1957, **114**, (6) 1220–1223). [In Russian]. By means of tracer isotopes placed in the materials of the charge and isotope detectors set in various positions in the walls of the furnace, it was possible to measure the rate of movement and the distribution of the batch inside the blast furnace. The wear of the lining was also studied with the aid of isotopes.—S. I. T.

Tapping Liquid Pig Iron from the Blast Furnace Hearth. B. L. Tavrog and E. D. Dmitrash. (*Metallurg*, 1957, (2), 2–3). A 1300 m^3 blast furnace built 1938 and reconditioned 1947 was put on high top pressure in 1952 and blown out in Jan. 1956. An account is given of the withdrawal of liquid pig iron through two holes, one 7 ft 2 in. below the axis of the iron notch on the same side and the other on the slag notch side 14 ft 4 in. below it. The upper hole yielded 656 t and the lower 391. This substantially reduced the time for repairs.

On the Properties at Elevated Temperature of Blast Furnace Brick. T. Hayashi. (*Tetsu to Hagane*, 1957, **43**, Mar., 266–269). [In Japanese]. Results are given for six types of brick, four imported, one made from domestic clay and one from domestic flint clay. They include refractoriness, porosity, air permeability, physical properties and abrasion volume loss in the cold; softening, expansion, shrinkage and reheating properties; and hot bending and abrasion data.

Study on the Behaviour of Compounds in the Blast Furnace. Making a Titanium Bear in the Laboratory and its Fusion. B. Eto. (*Tetsu to Hagane*, 1957, **43**, Mar., 211–213). [In Japanese]. Synthetic SiO_2–Al_2O_3–CaO–MgO slags containing up to 4·54% of TiO_2 were made up, and fused with pig iron, with oxygen additions, to simulate bears.—K. E. J.

Use of an Electric Integrator for Studying the Temperature Range in a Blast Furnace Hearth and in the Furnace Foundations. V. V. Yushkin. (*Stal'*, 1957, (9), 779–787). Conditions in hearths with a lower C layer as compared with other composite hearths and all-carbon bottoms are investigated. It is argued that previous statements that all-C hearths are best cooled round the circumference and from below may be incorrect. A design with a peripheral C lining is recommended though it has only small advantages over the fireclay hearth. The design causes concentration of heat at the top of the foundations near the carbon ring.—B. S.

Trials of the Partial Replacement of Metallurgical Coke by a Reducing Gas in the Blast Furnace. L. Naszalyi. (*Rev. Ind. Min.*, 1957, **39**, May, 423–435). An account of Hungarian experience is given. The role of CO in the furnace is outlined and the work of Raick and Brassert with natural gas is quoted and heat balances are indicated. Production and use of CO are then outlined and analyses of iron and slag are given.

Report on the Injection of Fuel Through the Tuyeres of a Blast Furnace. J. M. Tweedy. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, **14**, (2), 279–288). The results are given of trials carried out on a blast furnace by injecting fuel oil for a known time, through the tuyeres. The top gas was analysed before, during and after the trial. The injection of fuel did not have a significant effect on the furnace working. The hydrogen content of the top gas increased during fuel injection.—B. G. B.

Enrichment of the Blast and Injection of Fuel Oil into the Blast Furnace. J. M. Ridgion. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, **14**, (2), 289–295). The results are given of a study of the thermal requirements in the blast furnace and the influence of fuel oil and oxygen injected with the blast on the thermal distribution.—B. G. B.

Study of Mechanical Properties of Grey Cast Iron, Produced in Blast Furnaces with Charges containing Steel and Cast Iron Scrap. G. Savii. (*Met. Constr. Masini*, 1956, **8**, 57–61). The use of scrap is strongly recommended, not least because it represents a substantial economy. Charging of furnaces with up to 50% scrap is possible without special measures having to be taken. The only precautions would be to reduce by ~ 25% the proportion of the ordinary charges, to add a slightly greater quantity of coke so as to create an atmosphere of CO, 1% more will do, and to include a proportion of siliceous or cast iron ferro-silicon to compensate for the lack of Si. A detailed examination of the cast iron manufactured under good conditions shows its superiority to the ordinary product.—L. H.

The Centenary of the Cowper Stove. J. E. Lafon. (*Mét. Constr. Mécan.*, 1957, **89**, Sept., 749, 751–753, 755, 757, 758, 761, 763). A review of the development of the stove and various suggestions for its design or replacement are given.

On the Hot [Blast] Stoves at the Chiba Iron and Steel Works, Kawasaki Steel Corp. T. Nagai. (*Tetsu to Hagane*, 1957, **43**, Mar., 350–351). [In Japanese]. Dimensions and other details of the stoves are given with an account of Asawa-type checker bricks. A typical heat balance is set out.—K. E. J.

Study on Charcoal Pig Iron for Chilled Iron. III. K. Ohtani. (*Tetsu to Hagane*, 1957, **43**, Mar., 258–260). [In Japanese]. Specific heats and microstructures are compared for charcoal iron, normal pig iron and electrically-melted iron (analyses given), each with and without vacuum-melting treatment.—K. E. J.

The Economics of Remelting Czechoslovak Renn-process Balls in Blast, Low-Shaft, Reverberatory and Cupola Furnaces. A. Děkanovsky. (*Hutnické Listy*, 1957, **12**, (8), 673–682). Possible treatments of the Renn blooms containing P and S are reviewed. Cast iron, wrought iron or pig iron for the basic Bessemer process can be produced. It appears that blast furnace treatment is the most economical.

PROPERTIES, TREATMENT AND USE OF SLAGS

Dry Granulation of Blast Furnace Slags. R. Forbáth. (*Kohászati Lapok*, 1952, **7**, June, 121–127; July, 151–155). The characteristics of blast furnace slags are examined for their applicability to cement production. The effect of the cooling rate is investigated and critical temperature limits determined. A new, full-scale dry granulation method and the necessary equipment are described and the results reported. The properties of the granulated slag, are satisfactory for the production of cement.—P. K.

Basicity of Slags Containing Titanium Oxide. K. Mori. (*Tetsu to Hagane*, 1957, **43**, Mar., 380–381). [In Japanese].

A Study on the Rapid Photometric Analysis of Slags. T. Ikegami. (*Tetsu to Hagane*, 1957, **43**, Mar., 385). [In Japanese]. Details are given of methods for SiO_2, Al_2O_3, total iron, MnO, Cr_2O_3, P_2O_5 and TiO_2.—K. E. J.

[Rapid Determination of Basic Open-Hearth Slag Basicity, by pH and SiO_2 Determination]. I. H. Nakamura. (*Tetsu to Hagane*, 1957, **43**, Mar., 391–393). [In Japanese].—K. E. J.

Viscosity of Molten Slags of the System: Manganese Oxide–Silica–Alumina. Sh. M. Mikiashvili, A. M. Samarin and L. M. Tsylev. (*Izvest. Akad. Nauk, S.S.S.R. Otdelenie Tekn.*

Nauk, 1957, (1), 115–122). [In Russian]. The viscosity of molten slags of the system: $MnO–SiO_2–Al_2O_3$ corresponding in composition to the products of deoxidation of steel in the region of the lowest melting temperature was investigated. It was found that with increasing content of silica, the viscosity of homogeneously liquid slags increases and is determined mainly by the size of silicate anions. The viscosity of heterogeneously liquid slags at 1400–1590° C increases with decreasing silica content and is determined by the proportion of solid phase present. The lowest viscosity (0·5–10 poises) was observed for melts with $MnO : Al_2O_3 = 6$ and the content of silica from 20 to 30%. With decreasing $MnO : Al_2O_3$ ratio the viscosity of melts increases. An addition of alumina causes a slight increase in the viscosity of homogeneously liquid melts. The most fluid are melts with $MnO : SiO_2 = 2·6$ and containing up to 22% of alumina. With decreasing $MnO : SiO_2$ ratio, the viscosity of melts increases. An increase in the content of manganese oxide decreases the viscosity of homogeneously liquid slags . Melts of the lowest viscosity correspond in their composition to that of tefroite.

The Electrical Conductivity of Molten Slags Containing Titanium Oxide. IV. The $CaO–SiO_2–TiO_2$ System. K. Mori. (*Tetsu to Hagane*, 1957, **43**, Mar., 379–380). [In Japanese]. Changes in conductivity with CaO content (28–45 mol.-%) are shown for slags containing 19·5 and 30 mol.-%. For slags with a CaO/SiO_2 ratio of unity, changes in conductivity with TiO_2 concentrations in the range 18–48 mol.-% are shown.—K. E. J.

Transport Numbers of Iron Ions in Molten Iron Silicates. O. A. Esin and A. K. Kir'yanov. (*Izvest. Akad. Nauk, S.S.S.R. Otdelenie Tekh. Nauk*, 1956, (8), 20–27). [In Russian]. An electrolizer for the determination of transport numbers of iron cations in melts of the $FeO–SiO_2$ system was developed. As an indicator ^{59}Fe was used. It was established that transport numbers of iron cations in melts saturated with silica as well as those of calcium in the $CaO–SiO_2–Al_2O_3$ system are near to unity. In this way the cation character of the electrical conductiivty in the above melts and a low mobility of complex anions in such melts were confirmed. It was shown that in liquid slags of the system $FeO–Fe_2O_3–SiO_2$ the transport number of iron cations decreases continuously with increasing FeO content. This decrease is explained by two factors, namely by increasing electron conductivity and the participation in the transport of oxygen ions. An attempt was made to explain the predominantly electronic conductivity of $FeO–Fe_2O_3$ melts.

Blast-Furnace Slag. New Methods of Application in Road Construction. A. Genet. (*Arts et Manuf.*, 1957, Oct. 40–42). Crushed, granulated and expanded slags are briefly described and mixes used for road construction are indicated, with tables of their service life.

Use of Materials Based on Blast-Furnace Slag for Road-Making. Soc. Anon. Cockerill-Ougrée. (*Rev. Univ. Mines*, 1957, **13**, Aug. 15, 445–446). Use in tarmacadam is described.

WROUGHT IRON

Lockhart—Last Outpost of Hand Puddling. (*Iron Steel Eng.*, 1957, **34**, July, 160). A brief description is given of the last plant still producing hand puddled wrought iron in the U.S.A.—M. D. J. B.

DIRECT PROCESSES

By-Passing the Blast Furnace. E. L. van Devsen. (*Engineer Foundryman*, 1957, **22**, Aug., 36–39). A review of recent U.S. direct processes and sponge iron production is given in outline.

Economic Aspects of the Direct Reduction of Iron Ores. R. Touttée. (*Rev. Ind. Min.*, 1957, **39**, June, 547–567). Estimated tonnages of consumption and output are summarized with notes on prices, capital investments and trade balance. Technical processes are then reviewed with a general classification according to reducing conditions and physical state of the product. The position in the U.S., Germany, the U.K., Sweden and France is briefly indicated. Conditions for direct reduction in terms of ores, production capacities, saving in fuels and relative efficiencies and probable economic effects of direct reduction processes are then outlined.

PRODUCTION OF STEEL

On Fuel Consumption in Iron and Steel Plants. Y. Kuzuhara. (*Tetsu to Hagane*, 1957, **43**, Mar., 309–311). [In Japanese]. Equations are derived for fuel consumption at various stages in the steelmaking process, and nomograms are given from which fuel consumption can be estimated from operating conditions in a 1000-ton blast furnace, a 150-ton basic O.H. furnace and a soaking pit.—K. E. J.

Effect of Chromium on the Equilibrium of Carbon and Oxygen in Molten Iron Saturated with Carbon. III. S. Banya. (*Tetsu to Hagane*, 1957, **43**, Mar., 214–216). [In Japanese]. Results are obtained at 1400°, 1500° and 1600° C for Cr contents up to 19·6%, and are examined theoretically. The effect of Cr on the activity coefficient of O_2 is also deduced.

The Distribution of Sulphur Between Iron and Ferrous Slag. I. A. Tomilin and L. A. Shvartsman. (*Izvest. Akad. Nauk, S.S.S.R. Otdelenie Tekh. Nauk*, 1956, (10), 122–125). [In Russian]. The equilibrium distribution of S between iron and ferrous silicate slag was investigated using ^{35}S. The spread of results obtained was smaller than in investigations previously described, indicating that the method used was more accurate.—V. G.

Results of Measurement of Ventilation in Hot Shops. K. Maurer. (*Hutnické Listy*, 1957, **12**, (9), 808–813). A long-term examination of natural ventilation relating air exchange and temperature to external conditions is reported. Factors for keeping summer excess of working to outdoor temperature below 5° C are suggested.

Effect of Sulphur in the Furnace Atmosphere on the Hot-workability of Steel. T. Morishima. (*Tetsu to Hagane*, 1957, **43**, Mar., 256–258). [In Japanese]. An apparatus is described for heating steel specimens in atmospheres containing controlled amounts of SO_2. Results are presented showing the inuflence of SO_2 up to 0·4% in the atmospheres (containing varying amounts of CO or O_2) on the cracking of hot-bending test specimens of low-carbon steel heated at 1100° C for 1–3 h.—K. E. J.

Manganese–Oxygen Equilibrium in Liquid Iron. Y. Gunji. (*Tetsu to Hagane*, 1957, **43**, Mar., 218–219). [In Japanese]. Experimental results are given, from which are calculated the equilibrium between steam and H_2 and dissolved O_2 at 1550° and 1600° C for various samples, and at 1550° C for samples containing 3·83% and 3·24% Mn.—K. E. J.

On the Mn–S Equilibrium in Carbon-Saturated Molten Iron. III. Effect of Silicon. M. Inoue. (*Tetsu to Hagane*, 1957, **43**, Mar., 292–293). [In Japanese]. Data are given showing the effects of Si contents between $< 0·20\%$ and 8% on the Mn–S equilibrium in C-saturated Fe–C–Si melts at 1400° C, and the effects of Si on the solubility of C and the activity coeff. of S in the same melts. Desulphurization limits for Mn (0·5–3·0%) are calculated for Si contents up to 8%.

Removal of Copper from Molten Iron. M. Inoue. (*Tetsu to Hagane*, 1957, **43**, Mar., 265–266). [In Japanese]. Copper in amounts of approx. 0·75% is increasingly removed by additions of Na_2S or Na_2SO_4; 40% additions reduce the Cu content to 0·22% (71% removal of Cu). The content in the iron increases from 0·050% to 0·145–0·228%.—K. E. J.

The Influence of Sulphur On The Solubility of Oxygen in Liquid Iron. Z. Buzhek and A. Samarin. (*Doklady Akad. Nauk, S.S.S.R.*, 1957, **114**, (1), 97–98). [In Russian]. Experiments carried out at 1550° C and 1600° C show that S has no influence on the solubility of O_2 in liquid iron. This means that the solubility of O_2 in liquid iron as a function of temperature is expressed by the well known formula $\log[\%O] = -6320/T + 2·734$.—S. I. T.

The Facility Round-Out Programme at Columbia-Geneva. H. G. Degitz. (*Iron Steel Eng.*, 1957, **34**, Aug., 117–120). A description is given of the Geneva (Calif.) plant purchased from the U.S.A. government in 1946 by United States Steel Corp.—M. D. J. B.

Study of a Permanent Wall Type Converter with Compulsory Cooling. II. N. Tabata. (*Tetsu to Hagane*, 1957, **43**, Mar., 377–379). [In Japanese].—K. E. J.

On the Experimental Operation of a Turbo-Hearth. E. Honma. (*Tetsu to Hagane*, 1957, **43**, Mar., 221–223). [In Japanese].—K. E. J.

Influence of Oxygen of Low Purity on the Properties of Mild Steel. A. Ono. (*Tetsu to Hagane*, 1957, **43**, Mar., 381–383). [In Japanese].—K. E. J.

Rationalizing Basic Converter Practice. J. Marot. (*Silicates Indust.*, 1957, **22**, Sept., 468–474). [In French]. Recent work on the control of the Thomas type of converter is reviewed, and additional researches on that topic carried out in the Belgian Centre National de Recherches Métallurgiques are reported. The aim was to prepare the way for eventual complete automation. Variables suitable as control parameters are the bath temperature, the rate of blow of air, the total volume of air and variations in the flame spectrum during the blow.—P. F.

The United Steel Companies. C. Wilson. (*Steel Rev.*, 1957, July, 7–18). An illustrated survey.

Workington Iron and Steel Co. (*Brit. Steelmaker*, 1957, **23**, June, 168–173). An illustrated account of the plant and processes, with special reference to Railway materials, is given.

Ways of Developing Old Ural Works. D. D. Burdakov, M. I. Panfilov, I. P. Medvedev, D. P. Strugovshchikov, A. M. Nikolaev and K. V. Krasnov. (*Stal'*, 1956, (9), 818–820). [In Russian]. Examples of the low labour productivity and technological backwardness of some old iron and steel works in the Ural region are given and a 5-million rouble development scheme is outlined. This would increase labour productivity by factors of 5, 3·6, and 3·8 for pig-iron, steel and rolled products, respectively.—S. K.

Reconstruction of Old Ural Works Dams. V. S. Gvozdev and S. E. El'yasberg. (*Stal'*, 1956, (9), 831–835). [In Russian]. Many old dams and weirs are situated within works boundaries in the Ural region, dating back to the start of local iron making. The increasing importance of these water works and schemes for their reconstruction are discussed.

Recent Developments of Metallurgical Methods, with Special Reference to the Characteristics of Oxygen-blown Steelmaking. O. Cuscoleca. (*Tetsu to Hagane*, 1957, **43**, Feb., 162–166). [In Japanese]. A review is made of developments in America, Germany, France, Belgium, Holland, Austria, etc.—K. E. J.

Magnesium as a Desulphurizer for Steel Produced in Side-Blown Converters. B. A. Voinov. (*Liteĭnoe Proizvodstvo*, 1956, (3), 13). [In Russian]. The addition of a Si–Mg alloy to pig iron in the ladle before charging into a side-blown converter was found to lead to an average desulphurization of 34%.—S. K.

A Variety of the Russian Method for Producing Steel in a Converter. I. N. Ksintaris. (*Liteinoe Proizvodstvo*, 1956, (6), 29–30). [In Russian]. The author disputes the view that the "Russian" convertor process necessarily involves the use of low-silicon as well as superheated iron. A brief account is given of the operation of a 1·5-ton oxygen-blown converter with a charge containing 2·2–2·5% Si iron.—S. K.

Studies on the Oxidation of Phosphorus in Indian Pig Iron with Enriched Air. G. P. Chatterjee. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 117–125: discussion 126–127). The dephosphorization of Indian pig iron containing 0·3% P during top blowing with an oxygen-enriched blast was studied on a laboratory scale. The importance of the rate of rise of temperature under a given slag condition was confirmed. Although 0·06% P was the lowest obtained, it is considered that the process is worthy of pilot-plant scale trials.—A. D. H.

Effect of Oxygen Input Rates In the Decarburizing of Chromium Steel. G. W. Healy and D. C. Hilty. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, May, 695–707). The authors have developed methods for evaluating the heat balance of a furnace during O_2 blowing and estimating the change in Cr content, enabling the carbon content at the end of the blow to be calculated. The effects of O_2 input rate, Cr content, and bath temperature before blowing on metallic oxidation, O_2 usage, and final bath temperature, are also evaluated.—G. F.

Ways of Removing the Dust from Brown Converter Waste-Gases. W. Dehne. (*Stahl u. Eisen*, 1957, **77**, May 2, 553–562). Oxygen-blowing produces so much more brown waste gases than ordinary bottom-blowing that dust removal becomes necessary. The author lists quantities and compositions of waste gases under various conditions of blowing and of the various stages of blowing. He describes trials with a number of methods of dust removal, e.g. water-spray, steam injection, filtering, centrifuging and ultrasonic and electrostatic methods. The use of waste-heat boilers, which are highly efficient in the removal of dust, in conjunction with electrostatic filters was found to be the best solution.—T. G.

Roof Performance in the Third Campaign of an All-Basic Open Hearth Furnace. G. R. Rigby and H. M. Richardson. (*J. Brit. Ceram. Soc.*, 1957, **56**, Jan., 22–36). The third all-basic furnace at Stewarts and Lloyds, Bilston, Ltd., gave the record life of 2205 heats. An examination was made in the laboratory of (*a*) samples of the unused magnesite-chrome bricks, (*b*) a 4-in. thick peel from the working face of the roof after 633 heats, (*c*) bricks taken at the end of the campaign. Chemical, X-ray, and petrological examinations all revealed that iron oxide, lime, and possibly silica had been adsorbed at the hot face and had migrated into the brick. The iron oxide was confined to the first 3 in. but the lime had concentrated in a zone about 4 in. behind the working face. Deformation tests showed that all zones of the used brick were capable of sustaining a load at steel-making temperatures. Factors contributing to the long life of this roof are tentatively suggested.—D. L. C. P.

Door Jambs Made of Suspended Basic Refractory. (*Iron Steel Eng.*, 1957, **34**, May, 153). A brief description is given of a method of suspending refractories at O.H. front walls in the neighbourhood of the doors. The method was developed by the Geo. P. Reintjes Co. Reduced failures are claimed during furnace campaigns.—M. D. J. B.

Replacement of the Supporting Frameworks of a Tilting O.H. Furnace. V. I. Bodrov, M. G. Ermolov and V. I. Bogachev. (*Stal'*, 1956, (9), 845). [In Russian]. A brief account is given of the devices used and the organization of the work for changing the supporting framework of a tilting furnace during a ten-day cold-repair period. The whole furnace working structure (2000 tons) was lifted bodily with the aid of 16 jacks.—S. K.

An Evaluation of Cast Refractory Hot Top Linings. R. J. Tatousek and A. T. Peters. (*J. Met.*, 1957, **9**, Apr., Section 1, 582–585). Trials at Inland Steel Co. have shown that cast refractory hot-top linings are more costly and difficult to instal than brick linings, and although they last longer, excessive erosion of the cast lining during its extended life causes prohibitive hot-top metal losses. A 2·4% reduction in volume of the hot-top, obtained by increased taper, proved practicable.—G. F.

Bunker Method of Charging Granular Materials in Open-Hearth Furnaces. I. M. Krasinskiĭ. (*Stal'*, 1956, (9), 845–846). [In Russian]. The construction of a rail-borne bunker for supplying ore, limestone and other granular materials to the O.H. charging boxes is described. Two such bunkers have given excellent results at Magnitogorsk. The use of this system to save space and improve efficiency and productivity in the O.H. shop is recommended.—S. K.

Repair of the Smoke Stack of a Working Open Hearth Furnace. V. I. Morozov and M. T. Burnaev. (Stal', 1957, (1), 88–90). [In Russian]. O.H. stack repairs at a Soviet works take about ten days longer than furnace cold repairs. A water-cooled horizontal pipe connected to an induced-draught fan was successfully used as a temporary arrangement to fill the time gap, loss of steel production being thereby almost completely avoided.—S. K.

What Type of Chequers for Use in Open-Hearth Furnace Regenerators? J. Bečvář. (*Hutník*, 1956, **6**, (11) 341–343). [In Czech]. The design of regenerators and the choice of material for the bricks are discussed with special reference to regenerators working with O.H. furnaces having all-basic roofs. Results of experiments carried out in Czechoslovakia recently are given; the periclase-spinel type of brick appears to be most promising.—P. F.

Technical and Economic Characteristics of Recuperators. R. Kremer. (*Hutník*, 1956, **6**, (11), 324–329). [In Czech]. The various types of recuperators in use in the metallurgical industry are discussed from the point of view of thermal and economic efficiency, and the conditions under which the various types are most suitable are considered.—P. F.

Materials for Checkers in Open-Hearth Furnace Regenerators. V. A. Konopasevich. (*Stal'*, 1956, (9), 853–854). [In Russian]. The absorption of sulphur from producer gas by three types of refractory brick has been studied. The use of dolomite bricks, which were found to absorb sulphur satisfactorily, instead of forsterite for regenerator checkers is recommended.—S. K.

Advantages of Tilting over Stationary O.H. Furnaces. K. G. Trubin. (*Stal'*, 1956, (9), 780–782). [In Russian]. The author considers that the tilting O.H. furnace has many advantages and regrets its comparative neglect in the U.S.S.R.

where the wide diversity of raw materials should have strengthened the case for its adoption.—S. K.

Complex Automation of the O.H. Shop. G. A. Garbuz. (*Stal'*, 1956, (9), 782–789). [In Russian]. After a review of the state of O.H. automation in Soviet and foreign plants, Soviet practice is considered in detail. Schemes for linking the individually automated parts of the process into a complex system involving computers specially designed for the solution of equations involved are discussed. Some unsolved automation problems are considered, including control of evaporation-cooling and waste-heat boiler operation. Objectives recommended for research in the coming years include improved fuel-flow meters, rapid automatic steel analysis methods and the development of analytical expressions for the various parameters involved in steel melting.—S. K.

Automatic Control of Open-Hearth Furnaces. M. Gottwald. (*Hutník*, 1957, **7**, (1), 8–14). [In Czech]. Principles and equipment used in the automatic control are discussed. The best method at present seems to be based upon roof-temperature measurements coupled with the measurement of the oxygen-content of exhaust gases. Diagrams of the various schemes used for automatic control of O.H. furnaces are given. A comparison shows that in the Klement Kottwald Steelworks in Vítkovice an improvement in the lengths of campaigns, output and fuel consumption rates of the order of several per cent was obtained by conversion to automatic control.

Centralized Control of Power Systems in the Steel Industry. W. A. Derr and W. L. Metz. (*Trans. Amer. Inst. Elect. Eng.*, 1956, **75**, Part II, 332–339).

Electric Power Systems for Steel Plants. H. N. Cox and L. G. Levoy. (*Trans. Amer. Inst. Elect. Eng.*, 1956, **75**, Part II, 339–349).

Round Oak Steel Works, Limited. (*Brit. Steelmaker*, 1957, **23**, Aug., 236–238). An account on the occasion of the centenary.

General Layout of Metallurgical Plant. E. Kovařík. (*Hutnické Listy*, 1957, **12**, (5), 405–410). [In Czech]. The methodology and principles involved in the planning of integrated works, e.g. steelworks of 1·4 million tons p.a. capacity, are discussed.—P. F.

Cylindrical Converter of the Lenin Metallurgical Works (former Hungarian Iron and Steel Works Diósgyör). E. Weigl. (*Kohászati Lapok*, 1956, **11**, (8), 345–354). After referring to the historical development of the converting process especially in Hungary, the author comments on its present state in Diósgyör discussing in detail the iron conditioning and steel making experiments and the technology developed from them in the 18 ton cylindrical-horizontal converter of the Lenin Metallurgical Works. Afterwards he compares the durability of the various refractory linings. Finally he deals particularly with aspects and the economy of the method.—P. K.

Acid Bessemer Steel: Developments at Workington Iron & Steel Co. (*Iron Steel*, 1957, **30**, July, 357–360). Brief details are given of plant, processes, and production at Workington Iron & Steel Co., with emphasis on improvements in the past decade. The main points considered are the coal carbonizing plant, iron ore supplies and preparation, blast furnaces, acid Bessemer and electric steelmaking, rail and fishplate mills, and steel sleeper plant.—G. F.

Joint Investigations of the French I.R.S.I.D. and Institute of Welding on Steel Weldability. (1) Trend and Interest of the French Investigation on the Production of Basic Bessemer Steel; B.T.M. Steel (Steel with Low Metalloid Content). (*Soudage Techn. Connexes*, 1956, **10**, July-Aug.). A. Leroy, (184–185). Today there is co-operation between technicians, metallurgists and welders on the production of improved steels and the study of their weldability and serviceableness. An example is the basic Bessemer steel of low metalloid content (B.T.M.). The joint work of I.R.S.I.D. and the Institute of Welding has shown that this steel can be used in place of O.H. steel in certain welded constructions.

G. Husson. (186–191). Many welded constructions of O.H. rimming steel could be made of basic Bessemer steel provided the S, P and N contents are lowered and the scatter of these contents is lessened. Many methods of obtaining these objectives are described. The critical examination of the results obtained in experimental heats shows that one can obtain by a routine practice a basic Bessemer steel with S and P 0·035% max. and N 0·007% max. by observing a proper temperature control with the use of cooling additions and blowing with O_2 enriched air.

(2) Comparative study of experimental heats of extra-mild rimmed steels of O.H. and B.T.M. basic Bessemer processes. I—Mechanical Properties, R. Borione, F. Maratray and Y. Morillon. (191–199). Comparative tests on sheet steel of heats of O.H. rimming and Thomas BTM steel gave almost equal results for the transition temperature of the notch bar value in both the as finished and the aged conditions. The slight differences found could be explained by the chemical and micrographic examinations and in particular the fact that the C content of BTM steel is lower than that of O.H. steel. **II. Weldability tests, discussion and conclusion,** G. Granjon and J. P. Videau, (199–204). The Kinzel tests on weld samples in B T M and O.H. rimming steels show that, from the point of view of the sensitivity to notches, the weldability of the two steels is practically equal. For the heats tested, the Kinzel transition temperature of the basic Bessemer steel is lower and this appears to be due to the lower carbon content. The contents of O_2 and N_2 appear to act jointly on the results. The need to increase the C content of Thomas B T M steel in order to increase its resistance to fracture leads to the probability of there being a relation between the contents of C, O and Mn.

Deoxidation and Technological Properties of Killed Basic Bessemer Steels. I.—Deoxidation, Composition and Appearance of the Deoxidation Products. E. Plöckinger and R. Rosegger. (*Stahl u. Eisen*, 1957, **77**, May 30, 701–714). The trials were conducted in a conventional basic Bessemer converter; analysis of the iron used and the steel obtained are given. The course of deoxidation was followed by analysis of samples taken at frequent intervals. Deoxidation was by Si, Si + small amounts of Al and by Al. The reactions involved are shown in three-dimensional diagrams. The amount of non-metallic inclusions was found to be the same as that in O.H. steels. Al or a mixture of Al and Si were found best in keeping the amount of inclusions low. The degree of separation of the deoxidation products from the steel depends on the oxide phases formed as well as on temperature and the composition of the steel. **II.—Quantities of Non-Metallic Inclusions, Degree of Purity, and Mechanical and Technological Properties of Killed Basic Bessemer Steels.** (June 13, 798–804). It was found that separation of the deoxidation products in the ladle and mould is more complete the higher are the C and Mn contents. Tests showed that killed basic Bessemer steels having a low content of non-metallic inclusions in spite of high P and N levels of the melt reached strength values usually attained only with fine-grain O.H. steels.—T. G.

The Problem of Uniformity and Quality of Basic Bessemer Steel. P. Boutonnet. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, (5), 1003–1004). The author discusses this problem in general terms. M. Gombert. (1005–1030). A comprehensive review of factors influencing Thomas steelmaking is presented. An outline is given of research work carried out by I.R.S.I.D. on slopping in the convertor and its minimization by changing the convertor lines and on desiliconization. The need for regularity of the raw materials is stressed. Methods of controlling the actual blowing operation are discussed. A description is then given of the practice at the Société Métallurgique de Knutange and at three works of Sollac. The latter produce steel for tin plate and for deep drawing without the use of oxygen. Full operating data from these plants are given in graphical and tabular form. New blowing techniques are also described in which gases other than air are used in the converter.—B. G. B.

The High Dephosphorization of Basic-Bessemer Steel. H. Kosmider and H. Schenck. (*Stahl u. Eisen*, 1957, **77**, July 11, 917–926). High dephosphorization requires a slag of high lime basicity. Dephosphorization should be carried out with only small quantities of slag since loss of iron increases with increasing quantity. The P_2O_5 : P ratio of converter melts is smaller than that of melts treated in crucibles. High dephosphorization can be obtained by double slagging which is, however, not an economic proposition. The changes in slag composition as affected by mode of operation is shown by the ternary system FeO–CaO–P_2O_5. Sulphur in the melt is also reduced during the final dephosphorization treatment. Nitrogen pick-up can be avoided if O_2– enriched blast is employed. The results are presented in many graphs and equilibrium diagrams.—T. G.

The Development of Steel Making in Side-blown Converters in the Chinese People's Republic. Yu. Tszin' Shén. (*Stal'*, 1957, (7), 592–598). Side-blown convertors with nine tuyeres of 45 mm dia., a shell dia. of 1900 mm and 4000 mm high both of conventional shape and pear-shaped are described with details of operation; O_2 enrichment is used and also O_2 top blowing. They are basic lined and give steel of better quality than do bottom-blown converters.

Steel Refining with Pure Oxygen. O. Cuscolecs, K. Rösner and W. Kühnelt. (*Alpine Oxygen-Stahl, Mitteilung der Oesterreichisch-Alpine Montangesellschaft, Wien*, 1956, pp. 35). A comprehensive review of the Donawitz process in which the effects on C, P, Mn, S, N and O are carefully described. An account of the plant is given and the properties and uses of the steel are surveyed. (44 references).

Swedish Oxygen Steelmaking. F. Johansson. (*J. Met.*, 1957, **9**, July, Section 1, 972–975). An account of the Kal-Do process for high-P iron.

The Use of Oxygen in the Refining of Pig Iron. W. Bading. (*Stahl u. Eisen*, 1957, **77**, July 11, 926–931). A review on the use of oxygen either as pure gas or as O_2-enriched blast for the refining of iron in converters. The main part of the paper deals with safety precautions necessary in the use of large quantities of O_2.—T. G.

The Satisfactory Quality of High Carbon, Particularly Rail (Oxygen) L-D steels. H. v. Laizner. (*Radex-Rundschau*, 1956, (8), Dec., 384–397). After a survey of the production of high quality rimming steel by the L-D oxygen process, details of the properties of high carbon steels, including rail, patented wire and unalloyed tool steels for which the process has now been developed, are given. It is concluded that the quality is equal to that of high-quality O.H. steel. The possibility of making high-quality alloy steels by the process is discussed.

Roof Performance in the Third Campaign of an All-Basic O.H. Furnace. G. R. Rigby and H. M. Richardson. (*Refractories J.*, 1957, **33**, July, 308–315). An account of observations at Stewarts and Lloyds is briefly reported.

Measuring the Temperature at the Top of Gas Checkers in Open Hearth Furnaces at KMK. [Kuznetsk Works]. M. M. Epshtein. (*Stal'*, 1957, (7), 600–601). The lenses of radiation pyrometers become dirty from the gases so that they could not be used for this measurement. Pt-Rh–Pt and Chromel–Alumel couples were also unsatisfactory. An instrument which is cut off automatically during passage of gases has now been developed.

Automatic Control Applied to the Open-Hearth Furnace—Its Significance and Application. R. Alleyrac and P. Rodicq. (*Silicates Indust.*, 1957, **22**, April, 193–204). [In French]. A joint project involving collaboration between IRSID, the Wendel Steelworks at Hayange, and the instrument manufacturing firm MECI in Paris, for the automation of a 70-t O.H. furnace is described. Details are given of the fuel/air ratio and bath-temperature controls, etc. The mechanisms of the roof pressure meters, roof pyrometer circuits, gas analysers and other parts of the system are described. The uniformity of operation of the furnace attained by means of the system of automatic control resulted in a 10% increase in productivity and a 50% saving on refractories.—P. F.

Remelting High-Manganese Steel in the Open Hearth Furnace. P. T. Khokhlov. (*Stal'*, 1957, (2), 179–181). [In Russian]. Details are given of new practice for remelting 100% 11–14% Mn scrap in an O.H. furnace with reduction of Mn. Losses of Mn and hot-crack rejects have decreased since the adoption of the practice.—S. K.

The Effect of Manganese on the Rates of Desulphurisation and Oxidation of Metal in the Open-Hearth Process. V. Friedrich. (*Hutnické Listy*, 1957, **12**, (4), 324–329). [In Czech]. The chemistry and thermodynamics of Mn reactions in the O.H. furnace are discussed. It is shown that high Mn contents are beneficial in steels whether these have been made by the ore or by the scrap processes. Mn reduces the oxidation of the metal and improves its impact strength after ageing. The range of Mn contents studied was 0·1–2·0%; the steels were of the basic type.—P. F.

Opening Open-Hearth Furnace Tapping Holes with Explosives. M. E. Savos'kin. (*Stal*, 1957, 181–182). [In Russian]. Experience at the "Zaporozhstal" works on the adoption of the explosive method used at Czech works for the rapid opening of O.H. tapping holes is described. Results have been encouraging and wide adoption of the practice is recommended.—S. K.

Change in the Hydrogen Content of Open-Hearth Steel During Melting. A. L. Bogorodskii and V. A. Noskova. (*Stal,'* 1957, (2), 182–183). [In Russian]. The hydrogen content of the metal at various stages of the process in basic and in acid O.H. furnaces has been studied on the basis of analysis of 600 samples for over fifty heats. Results show discrepancies from those of other investigators.—S. K.

Hot (Evaporative) Cooling of Open Hearth Furnaces. S. I. Moiseevich. (*Stal'*, 1957, (7), 658–662). Further advances should be concerned with reduction of heat loss and increased use of steam. The process as used as the Makeyevka plant is described and details of heat losses from the cooling of ports, skewback channels, valves and dampers are given. The use of chrome-magnesite reduces this by 25–50% in the ports and by 40–80% in skewback channels. The total reduction for a 350-t furnace can be better than 30–35%.

Water Practice for Open Hearth Furnace Evaporation Cooling Systems. A. P. Mamet, A. V. Nikolaev and A. I. Kabanova. (*Stal'*, 1957, (2), 173–178). [In Russian]. The results are presented of a comprehensive study of water practice and thermal and chemical investigations of evaporation cooling systems at two O.H. shops at different works. These results have been used for establishing temporary standards for such systems.—S. K.

Method of Testing Open Hearth Precipitators. J. H. Smith and G. L. Rounds. (*Iron Steel Eng.*, 1957, **34**, June, 139–141). This paper describes the methods and equipment used to test the efficiency of a precipitator installed at an O.H. furnace at Kaiser Steel Corp.—M. D. J. B.

Design and Construction of Fontana Open Hearth Precipitators. E. V. Akelow. (*Iron Steel Eng.*, 1957, **34**, June, 131–138). This paper deals with some of the problems of design and construction of an electrostatic Cottrel type precipitator installed at one of the Kaiser O.H. furnaces.

Properties of Open-Hearth Steel Melted with the Use of Oxygen. D. S. Kazarnovskii, T. M. Ravitskaya, M. P. Sidel'kovskii and L. P. Tarasova. (*Stal'*, 1957, (2), 152–157). [In Russian]. The results of determinations of mechanical properties and chemical compositions of O.H. rail steel (0·6–0·8%C) produced with and without the use of oxygen are presented and discussed. Both special experimental results and an analysis of four years' routine test data are included. The steel was made from a high-P iron. No deleterious effect on the mechanical properties studied were produced by flame oxygenation before deoxidation or bath blowing during finishing. The static strength, plasticity or fatigue strength of carbon steel oxygen blown in the bath to within 55–8 min of the start of deoxidation did not suffer; but some decrease in toughness and increases in O_2 content and fatigue rate occurred in steels when the blowing was finished less than 60 min before deoxidation.—S. K.

The Use of Oxygen to Further the Thermal Efficiency in Open Hearth Furnaces. O. I. Jacunszkaja and M. N. Sztarovics. (*Kohászati Lapok*, 1953, **8**, July, 151–162). A short review is given of the development of oxy-oil burners in the U.S. It is suggested that O_2 blowing should begin after charging $\frac{1}{3}$ or $\frac{1}{4}$ of the scrap. The use of O_2 for melting a hot charge of more than 65% is not favourable. The purity of O_2 should be 90% or more, although the results of experiments using only 75% O_2 were also satisfactory. Some advantages of using O_2 are discussed; these include high temperature, decrease in oxidizing ore, lime and slag, reduction of sulphur absorption and increase in steel output.

A Combined Method Using Oxygen in the Open Hearth Steelmaking Process. O. I. Jacunszkaja and M. N. Sztarovics. (*Kohászati Lapok*, 1953, **8**, Dec., 258–260). The substance of this method is the use of oxygen in end burners during the charging and melting periods and the blowing of oxygen into the bath during the finishing period. During a 200 ton. experimental cast at the Republic Steel Corporation 6·8 m^3/ton oxygen was used in the end burners and 17·1 m^3/ton was blown into the bath. The higher temperatures achieved by this method facilitates slag removal. According to data of the Bethlehem Steel Corporation using this method, the time of the heat was reduced to 50% without spoiling the quality of steel, but simultaneously the repairing time of the furnace increased by 42%.—P. K.

Oxygen Conversion in the Open Hearth. (*Steel*, 1957, **140**, June 3, 92–94). Experiments and production trials are

described in which quality carbon steel was produced with O_2 and no external heat. After tests, 4 125 ton heats were made in an O.H. furnace. Cold metal, hot metal and additions were put in the furnace and 99·5% O_2 injected through three lances at the slag-metal interface. Resulting steel composition was good with high C and low P, S and N. Refractory wear may be troublesome, but a specially designed reactor vessel is foreseen.—D. L. C. P.

Furnace Pressure and Heat Transfer. C. H. Stone. (*Brit. Steelmaker*, 1957, **23**, Aug., 246–249). A study of the O.H. furnace is described. Thermocouples were inserted in the roof, back wall and 3 ft from the tap hole before warming up and records for a complete charge are shown. The pressure record is also reproduced and preliminary findings are given. It is confirmed that a buoyancy pressure difference of 0·012–0·013 in w.g./ft height occurs in the furnace. High pressure prevents roof overheating but roof temp. is not a criterion of rate of heat transfer to the bath.

Experience in the Conversion of Open-Hearth Furnaces to Natural-Gas Firing. V. P. Borodin, P. E. Darmanyan, I. A. Yudson and L. S. Shevandina. (*Stal'*, 1957, (2), 124–129). [In Russian]. After a brief note on the economics and probable development of the use of natural gas in Soviet steel works, details are given of the operation of one basic and one acid-roof small (about 17·5 m^2 bottom area) O.H. furnaces fired with a natural gas/oil mixture at the "Krasnyi Oktyabr'" works. The first produced mainly carbon steel, the second ball-bearing steel; optimal conditions for the new practice, which is especially advantageous in producing quality and high-quality steels, are summarized. Data are compared with those obtained during oil-firing.—S. K.

The Use of Rhébinol Mortar in the Open Hearth Plant at Hagondange. (*Centre Doc. Sidér. Circ. Inform. Tech.*, 1957, (5), 1036–1038). A short account of the use of Rhébinol mortar, which is based on zirconium silicate, for repair of O.H. brickwork is given.—B. G. B.

The Reduction of the Melting Loss of Deoxidizing Materials and Indigenous Inclusions in Open Hearth Steels. D. Csépai. (*Kohászati Lapok*, 1955, **10**, May 211–217; June, 257–262). The main factors of the increase in yield of Mn in the O.H. process are as follows: (1) Slag ratio about 2·2, (2) minimum quantity of slag, (3) high melting temperature, (4) Predeoxidation with 45% ferrosilicon or ferromanganese silicon before the addition of ferromanganese. Besides these factors, (5) slag removal after the melting period, (6) high tapping temperature and (7) the prevention of mixing tapped steel and slag are also important for the reduction of indigenous inclusions.—P. K.

Preparation and Properties of Boron-Treated Non-ageing Open Hearth Steel. E. R. Morgan and J. C. Shyne. (*J. Met.*, 1957, **9**, June, 781–785). Addition of >0·005% B as ferroboron seriously affects rimming characteristics, 0·007% is required to avoid strain ageing without supressing rimming, and it is best to use at least 0·004% B as a capping addition. This is most efficient when added before the Al cap. Nitrogen strain ageing is completely eliminated and ingot-to-slab yield of 85% can be combined with increased ductility as shown in deep-drawing tests.

Observations of Slag and Flue Dust Deposits in Under Part of an Open-Hearth Furnace. E. Neumann and J. Seitz. (*Radex-Rundschau*, 1956, (8), Dec., 377–383). Observations of slag and flue dust deposits in the slag pockets, checkers, flues and stack of an O.H. furnace system are described for acid and basic construction and for oil and gas firing. A change from gas to oil firing resulted in an increase in the flue dust deposit but the amount of slag deposited in the pockets was unchanged. The dust deposits exercise a determining influence on furnace life. Investigations of the slag encrustations on fireclay checker bricks showed that those of higher fusion point were associated with less checker wear than the others. The chemical and physical nature of deposits from various parts of the sub-hearth system are given and reasons for the differences are advanced.

Method Used During Forced Interruptions in Open Hearth Operations. G. A. Ferris. (*J. Met.*, 1957, **9**, Apr., Section 1, 591). From a questionnaire survey of U.S. and Canadian O.H. plants, brief details are given of four major methods used in shutting down O.H. furnaces for known or varying periods of time.—G. F.

Desulphurization Tests on Steel. (*Canad. Mines Tech. Surveys Phys. Met. Division*, (*July-Dec.* 1956), 1957, Jan. 15, 1–2). Quicklime and Al–Mg alloy were used on 500 lb melts in an acid-lined ladle at 1700°.

Hydrogen in Steelmaking Practice. H. Epstein, J. Chipman, and N. J. Grant. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 1, 597–608). A series of 3 acid O.H., 2 basic O.H., and 7 basic electric heats have been sampled and analysed for H_2 at stages during refining, at tap, and during pouring. The measured H_2 content is shown to be the result of two competing processes, one its absorption from gas and slag phases and the other its removal during the CO boil. H_2 absorption rates are calculated, showing the differences between the three steelmaking processes.—G. F.

Open Hearth Furnace Tapholes Placed by Gun. (*Indust. Heating*, 1957, **24**, Feb., 364–365). The closing of O.H. tapholes by a mud gun has led to more efficient operation and reduction in delay.—A. D. H.

Explosive Lances (Jet Tappers) for Opening Tapping Holes of Open-Hearth Furnaces. I. The Principle of the Explosive Lance. II. Experience Gained with Explosive Lances. (I) J. Prior. (II) W. Burmeister. (*Stahl u. Eisen*, 1957, **77**, May 2, 562–563; 563–567). [I] The principles of the jet tapper are explained. [II] Design and use of the jet tapper is described and experience gained in the opening of tapping holes of O.H. furnaces discussed. The costs and savings in time and money are analysed. In the discussion, the various reasons for unsuccessful results from the use of jet tappers are pointed out.—T. G.

Distribution of the Metal in the Ladles during Tapping in a Branched Runner. M. K. Tkachenko. (*Stal'*, 1956, (9), 844). [In Russian]. Losses due to incorrect distribution of the flow of steel from an O.H. furnace between the two branches of a runner system are briefly described. Devices for controlling the distribution, including a new design of gate brick which minimized the pressure of the metal, are discussed.

On the Oxidation of Liquid Steel during and after Tapping from an Open-Hearth Furnace. L. S. Gorokhov and I. F. Minchenkov. (*Stal'*, 1956, (10), 934–935). [In Russian]. The authors consider that the oxidation of steel during tapping occurs in the stream and not in the ladle and suggest experiments to decide the extent of oxidation inside and outside the furnace. They consider many of N. S. Mikhailets' recent statements on this subject misleading.—S. K.

On the Oxidation of Liquid Steel Outside the Furnace. N. S. Mikhailets. (*Stal'*, 1957, (6), 520).

Cleaning Hot Gases. L. Silverman. (*Ind. Eng. Chem.*, 1957, **49**, July, 67A–68A). An account of slag wool filters, as used for O.H. fume removal, is given. (4 references).

The Hot-Blast Cupola in Open-Hearth Steel Plant. A.-F. Oberhofer. (*Stahl u. Eisen*, 1957, **77**, May 16, 643–651). It is more economical, also from the thermodynamical point of view, to melt down scrap for the O.H. plant in a hot-blast cupola than to instal a new O.H. furnace. Good desulphurization is achieved by melting under a highly basic slag. Mode of operation, the reactions involved, etc., are described.

Electric Smelting of Blast Furnace Flue Dust in the 60KvA Arc Furnace. (*Canad. Mines Tech. Surveys, Research and Special Projects*, (*July-Dec.* 1956), 1957, Feb. 8, 17–18). Particle size and composition were determined and smelting trials were carried out. Pelletization with lime and water is satisfactory and prevents dust loss.

Use of Steel Shavings in Electric Steel Melting. N. I. Domrachev. (*Liteinoe Proizvodstvo*, 1956, (2), 22–23). [In Russian]. A brief account is given of acid electric-furnace practice with increased (up to 1130 kg. of a total charge of 3487 kg) utilization of steel shavings. Data on steel compositions and properties corresponding to various properties of shavings are shown.—S. K.

Production of Carbon Steel in Acid Electric Furnaces by the Silicon-Reduction Process under a Fireclay Slag. A. I. Shvedunov and E. I. Lopatkin. (*Liteinoe Proizvodstvo*, 1956, (2), 5–7). [In Russian]. A process developed and introduced in 1954 is described in which carbon steel is made in electric-arc furnaces, old firebrick being used instead of quartz sand for forming the slag (typical composition: 50·3% SiO_2, 26·2% Al_2O_3, 8·36% Fe_2O_3, 2·62% FeO, 1·6% CaO, 10·24% MnO) during melt-down. The fireclay slag has a low electrical conductivity, leading to rapid heating of the bath and reduction of silicon to 0·20–0·50%. The steel is highly deoxidized and degassed. The alumina

content of the slag obviated the normal difficulties of producing high-manganese steel in acid electric furnaces. Full details for working the new process are given.—S. K.

Efco Graphite Resistor Furnace. (*Metalworking Prod.*, 1957, **101**, May 24, 890). A 15-cwt. furnace for carbon and low-alloy steels is described with power rating 350 kVA. The roof is removable and so is the side door improving accessibility.

Calculation of the Diameter of Self-Sintering Electrodes for Ore Reducing Electric Furnaces. E. M. Alekseev. (*Stal'*, 1956, (10), 894–896). [In Russian]. It is shown that for Soderberg type electrodes the product of the optimal current density and the cube root of the electrode material consumed per unit of electrical energy is constant. Thus, if the optimal value for producing one alloy is known, that for any other can be calculated.—S. K.

Use of Oxygen in Electric Steel Production. N. I. Shutkin. (*Stal'*, 1956, (10), 890–894). [In Russian]. In 1950, 5056 tonnes of electric-furnace steel were produced at the "Elektrostal'" works with oxygen blowing, in 1955 the figure had risen to 36,474. This article reviews two variants of the practice for a wide range of high-alloy stainless, tool and structural steels. Oxygen blowing accelerated the process, reduced electricity consumption and as a rule, gave a better quality product with less rejects. The practice evolved avoids excessive lining wear and oxidation of chromium; it is exemplified by details for a heat of 1Kh18N9T steel (C $\leq$ 0·12%; Si $\leq$ 0·8%; Mn $\leq$ 2·0%; S $\leq$ 0·030%; P $\leq$ 0·035%; Cr, 17–20%; Ni, 9·0–11·0%; Ti, 0·5–0·8%.—S. K.

Desulphurization of Steel by Soda Ash Treatment. N. J. Wadia, V. G. Paranjpe, and S. Visvanathan. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 109–115; discussion 115–116). Production-scale trials showed that low carbon steel could be desulphurized by soda ash only after deoxidation. Calcium silicide was a much more efficient desulphurizer for 4% Si steel and the efficiency of soda ash decreased with increasing oxygen potential of the sulphur-containing melt.

Influence of Deoxidation with Aluminium on Structural Steel Quality. N. G. Antropova, Z. M. Kalinina and A. K. Petrov. (*Stal'*, 1957, (1), 64–69). [In Russian]. In the investigation described the influence of deoxidation with varying additions of Al to O.H. and electric-furnace steel on the structure, fracture and mechanical properties was studied. From the results obtained and information in the literature the following explanation is advanced of the previously reported and newly confirmed important influence of Al and N_2 on alloy structural steel properties: only part of the Al added for final deoxidation is oxidized, the rest appearing at the grain boundaries on crystallization. Rapidly diffusing N_2 combines preferentially with this to give a very stable and harmful aluminium nitride envelope round the primary grains.

Rimming Steel Deoxidation Practice with Ferromanganese in the Ladle. I. I. Bornatskiĭ, K. S. Alferov, and V. I. Prilepskiĭ. (*Stal'*, 1956, (11), 977–983). [In Russian]. Tests at Makeevka on the ladle deoxidation of rimming steel (produced in 185–370-ton basic-roof O.H. furnaces on 65–70% liquid metal scrap-ore process) showed that uniform distribution of Mn in the liquid steel was obtained with a sufficiently high pre-tapping temperature. Compared with deoxidation in the furnace an average ferromanganese saving of 25% was obtainable. The new practice resulted in no significant changes in the behaviour and properties of the steel in the liquid or solid states. A maximum ferromanganese lump-size of 80 mm is recommended. Wide adoption of the practice is recommended for basic-roof furnaces.—S. K.

Suggestions for the Improvement of the Quality of Our Steels. W. Küntscher. (*Neue Hütte*, 1957, **2**, Feb.-Mar., 154–157). Slag composition should be so controlled as to improve the permeability to gas of the molten slag. Casting at negative pressures and in protective gas atmospheres is discussed. A new development of the Perrin process, the so-called "reduction alloy process," which is particularly suitable for steel production in large electric furnaces, is described. Modern methods of heat treatment are surveyed.

Machine-Construction Problems in the Field of Open-Hearth Furnace Equipment Development. V. A. Brandt. (*Stal'*, 1957, (2), 119–124). [In Russian]. The possibilities of making available to Soviet O.H. furnace shops improved types and sizes of the following equipment are discussed:—charging machines; cranes; hot metal mixers; bogies for charging boxes and ingot moulds; hot metal pouring and slag ladles; valve reversal gear. Further mechanization is also considered.—S. K.

Basic Hot Blast Cupola as Source of Hot Metal for Steel Plants. E. S. Harman and S. Tunder. (*Iron Steel Eng.*, 1957, **34**, June, 159–163). This paper calls attention to the possibilities of the development of the hot blast cupola and describes its advantages over the cold blast acid-lined cupolas installed in a number of steel plants in the past. Hot blast cupolas provide a means of getting hot metal for steelmaking without the costly expenditure of additional blast furnaces. Operational and economic data are given.—M. D. J. B.

Steelmaking Technology in Acid Lined Arc Furnaces. A. Toth. (*Kohászati Lapok*, 1954, **9**, June, 130–137). The construction of an acid lined 3 ton Héroult furnace, the steelmaking technology and the economy of this process, especially in the steel foundry, are discussed in detail.—P. K.

Arc-Furnace Studies. B.I.S.R.A. (*Foundry Trade J.*, 1957, **103**, July 25, 111–112). Power input, electric power and electrode consumption were studied on an experimental furnace of modified Héroult type. Roof wear is also briefly noted.

Experience in the Reconstruction of a Steel-Melting Arc Furnace. A. I. Kholodov and S. I. Suchil'nikov. (*Stal'*, 1957, (2), 184). [In Russian]. Replacement of the cylindrical shell of an arc furnace by a conical one and other constructional changes were found to give greatly increased lining life without loss of productivity or specific power consumption.

Experimental Results with Hollow Electrodes in Electric Steel Furnaces. W. E. Schwabe. (*Iron Steel Eng.*, 1957, **34**, June, 85–91). The paper describes a programme of laboratory and field tests carried out to study the electric furnace arc, the conditions of its erratic behaviour during melt down and methods of improving arc steadiness.

Regulating Systems in Electric Arc Furnaces. J. G. Robin. (*Mét. Constr. Mécan.*, 1957, **89**, Apr., 359–363). The importance of regulation in arc furnaces is first described. Regulating systems are based on one of three principles: constant current, constant voltage or constant impedance. Modern arc furnaces use constant impedance. A number of commercial units are described.—B. G. B.

Processing of Cr-Ni-W Steel Scrap in the Electric Furnace. K. Hýbek. (*Hutnické Listy*, 1957, **12**, (4), 377–384). [In Czech]. Details are given of works experiments designed to make pre-alloy for the production of Cr-W-Ni steels from O_2-refined scrap, mostly shavings, obtained from the finishing process of Cr-W-Ni steels. The utilization of the pre-alloys was shown to be possible on the basis of test melts carried out in 0·5-t arc and 1-t high-frequency steelmaking furnaces. Details of the preparation of the pre-alloy, the method of its utilization, as well as results of mechanical tests and metallurgical studies of the steels (CrNiW–22–11–3) made with the pre-alloy are given.—F. F.

Analogue and Controlling Computers for Electric Arc Furnaces. Yu. E. Efroimovich. (*Stal'*, 1957, (7), 602–608). The perfecting of methods of control of electrical conditions is described. Good results were obtained both from the points of view of economy and of steel quality.

Stainless Steel in Britain. P. Slater. (*Metal Bull.*, 1957, July, Special Issue, 19–26). A review of uses with a table of producers.

French Stainless Steel. M. Maroy. (*Metal Bull.*, 1957, July, Special Issue, 47, 49, 51–52). **Stainless Steel in Sweden.** E. Öhman, (55, 57, 61). **Stainless Steel in Japan.** (63, 65, 67). Notes on production and types with lists of producers are given.

The Stainless Steel Industry in the U.S.A., R. E. Paret. (*Metal Bull.*, 1957, July, Special Issue, 29–45). A brief review with numerous tables of standard types, production, shipments by market classifications, and producers. Canada is included.

Weldable and High Strength Special Steels for Steel Structures. A. Legat. (*Schweisstechn.*, 1957, **11**, April, 45–48). The weldability of high strength structural steels and the effect of deoxidation method on the physical properties of the steel are discussed. The behaviour of "St 37" at low temperatures for air separation units and weldability are discussed.

Ultra-High Strength Steel. (*Nat. Bur. Standards Tech. Bull.*, 1957, **41**, June, 81–82). Properties of steels based on A.I.S.I. 4340 are briefly indicated. They were designed for high-resistance as well as high tensile strength.

Design Properties of 17-7 PH Stainless Steels. (*Machine Design*, 1957, **29**, Apr. 4, 138). Sheet, bar and wire in annealed-air cooled, cold-drawn, and cold-drawn and hardened at 900° F conditions are available and these can be further heat treated to condition T and condition TH 1050 or to an ultimate tensile strength of 244000–345000 psi in wire form. A dimensional increase occurs. Any heat tint is a potential source of rust which has a passivating effect but shows pits on removal.

Avesta 664 Stainless Steel—A new Austenitic Chrome-Nickel-Manganese Steel. (*Acciaio Inossid.*, 1957, **24**, March-June, 61–67). [In Italian]. A new stainless steel developed by the Avesta Steelworks is described; the steel has 0·08% C, 17% Cr, 4·5% Ni, 6·5% Mn and 0·15% N. Its mechanical, physical and anticorrosive properties are described as well as its workability and weldability.—M. D. J. B.

Boron Solves "Hot Shortness" in Stainless Steels. D. L. Loveless and F. K. Bloom. (*Iron Age*, 1957, **179**, June 20, 95–97). It is reported that small boron additions in the ladle improve the yield of austenitic stainless mill products by greatly improving hot working performance; also less reheating and grinding operations to remove surface defects are necessary. Additions are normally <0·005% B.—D. L. C. P.

New Stainless Steels Containing 17% Cr–4% Ni–6% Mn and 18% Cr–5% Ni–8% Mn. (*Amer. Iron Steel Inst., Contrib. Met. Steel*, 1955, No. 47, pp. 9). Shortage of Ni has led to the development of stainless steels in which Mn replaces part of the Ni in 17/7 and 18/8 steels. The new 17/4/6 composition is a satisfactory substitute in some applications where machinability and severe forming characteristics are not essential: where they are essential, the 18/5/8 composition is more appropriate. Details are given of the mechanical and physical properties, corrosion resistance, and fabrication characteristics of the new steels.—G. F.

Investigation on the Development of Low-Carbon Low-Alloy Case Hardening and High Strength Steels with Electrolytic Manganese. S Visvanathan, S. N. Anant Narayan and P. K. Ohakravarty. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 187–214; discussion 214–216). The possibility of using electrolytic Mn in the production of case-hardening and weldable high tensile steels is discussed. The properties and heat-treatment of 0·5/2·0% Mn, 0–1·2% Cr steels were determined and several compositions proved satisfactory for case hardening. Steels containing 1·2–1·6% Mn, 0·1% Ti, and 0·3% Si, had physical properties comparable with those of nickel-bearing structural steels.—A. D. H.

A New Method of Deoxidation and Desulphurization of Steel with Improvement in Quality. V. A. Skachks and N. P. Merenkov. (*Stal'*, 1957, (6), 521–522). Deoxidiser tubes of steel containing molten Al were compared with rods in steel treated with ferromanganese in the ladle. Steel from the tube method had greater ductility and impact and tensile strength. Nitride formation with production of fine grained structure is advanced as the cause. Both C and S were reduced and inclusions were spheroidal and within the grains. It is recommended as a substitute for low alloy steels, e.g. in reinforcement for concrete.

Progress in Stainless and Heat Resisting Steels. J. Lomas. (*Eng. Boiler House Rev.*, 1957, **72**, Feb., 56–57). Recent developments in metal grades and processes of interest to the engineer are dealt with.—D. L. C. P.

Stainless Steel. A Review of Its Advantages and Uses. P. M. Slater. (*Sheet Metal Ind.*, 1957, **34**, July, 503–516). A review of the use of sheet and strip in the automobile, hospital equipment and food handling, and building industries and in cutlery and kitchenware, with a note on aircraft is given.

Stainless and Heat Resisting Steels. (*Amer. Iron Steel Inst. Steel Prod. Man.*, 1957, June, pp. 90). A list of commercial grades and the bars, wires, etc. made from them with their properties and other data.

Nickel-Free Austenitic Stainless Steel. (*Mat. Methods*, 1957, **45**, Jan., 104–105). Mechanical, welding and corrosion properties of "Tenelon" (C. 0·1% max., N 0·4%, Cr 17%, Mn 14·5%) are given and compared with nickel-bearing stainless steels.—A. D. H.

High-Temperature Stainless. Eight Alloys for 600–1400° F. E. A. Loria. (*Product Eng.*, 1957, **28**, Apr., 135–139). Compositions and mechanical properties of austenitic, semi-austenitic and martensitic steels are given with notes on their fabrication. The alloys are HNM, HTX, and A-286; AM-350 and 17–7PH; and 420, 422, and 422M steels respectively.

Castable Alloy for Highly Stressed Parts. W. C. Clarke. (*Precision Met. Mold.*, 1957, **15**, April, 41–42, 106). Armco 17–4 PH, an age hardenable corrosion-resistant steel, is described.

Influence of Technological Factors on the Contamination of Type ShKh15 Steel. M. I. Zuev, D. P. Zhuravskiĭ, M. I. Vinograd, and M. A. Lyubinskaya. (*Stal'*, 1957, (1), 43–47). [In Russian]. An account is given of the statistical treatment of quality control and operating data on oxide inclusions in a ball-bearing steel and comparison with results of special experiments. The steel is produced in 20-t arc furnaces by remelting, by using a fresh charge or by a duplex process. Minimal oxide inclusions and hair-line cracks are obtainable when furnace lining was in good condition and metal temperature in the ladle (with a high-stability lining) was kept at 1530–1580° C; when 500 kg ingots are filled in 80–100 seconds on an 8-ingot stool. Better organization of stool and ingot preparation work is also recommended.—S. K.

Some Defects of 38 KhMYuA Steel. A. D. Kramarov, A. M. Levin, N. V. Tolstoguzov, and K. M. Konovalov. (*Stal'*, 1956, (11), 1015–1021). [In Russian]. There has been considerable discussion in recent years on the qualities of nitrided type 38KhMYuA steel, which contains about 1% Al, and in this article extensive results of process control tests and special experiments are analysed. The mechanisms through which the main defects (hair-line cracks, spotty segregation, flaky fracture) arise have been deduced and methods for avoiding them specified. In their nature and prevention the defects resemble those encountered in several other steels, but are modified by the high Al content of the present steel.—S. K.

Properties of 20 KhGR Steel. N. I. Letchford. (*Stal'*, 1957, (6), 564). A boron-containing steel for the automobile industry is described. It contains 0·17–0·24% C, 0·17–0·37% Si, 0·7–1% Mn, < 0·04 S and P, 0·8–1·1% Cr, 0·3% Ni and 0·002–0·005% B. Ti is used for deoxidation before adding B. Mechanical properties are given and the hardenability diagram is shown.

The Structural Stability of Alloys Having High Chromium and Carbon Contents. K. Löbl, H. Tůma, and J. Ježek. (*Hutnické Listy*, 1956, **11**, (10), 592–599). [In Czech]. The influence of annealing temperatures and times on the structures formed in Fe–Cr–C alloys containing 1·5–4·5% C and up to 31% Cr was studied by means of chemical, electron-micrographic and X-ray methods. Annealing times reached up to 1000 hours annealing temperatures were 550° and 1000° C. The structural changes occurring in the alloys, and the carbides identified are described.—P. F.

Manufacture of Large Ingot Moulds. Avoidance of "Corner Scale." W. S. Williams. (*Iron Coal Trades Rev.*, 1957, **175**, July 12, 104). Corner scale has been found to be high in P and must represent low melting metal forced out by pressure. Prevention is possible by supplying artificial hot spots which can be done by blowing the top of the mould with air through a tube between the pourer head and riser head. A rough formula for the strength of blow necessary is given.

Material for Steelworks Moulds and Some Aspects of Processing. J. Přibyl and B. Zeman. (*Hutník*, 1956, **6**, (10), 296–299). [In Czech]. Grey cast iron is recommended, and the method of mould manufacture considered to give best results is discussed.—P. F.

Some Aspects of Steelworks Mould Construction. J. Přibyl. (*Sborník Vědeckých Prací Vysoké Školy Báňské v Ostravě*, 1956, **2**, (2), 101–105). [In Czech]. The inadequacy of the criterion of the ratio of the ingot weight to that of the mould weight in mould design is pointed out, and the relatively greater importance of wall thickness and absence of discontinuities in the walls is emphasized. The life of moulds is shortened by thermal shock, the latter being severe with ingots having concave or wavy profiles. The author proposes abolition of such shapes.—P. F.

Durability of Ingot Moulds for Steel Pouring. S. M. Bobrovskikh and A. G. Nikolaev. (*Stal'*, 1957, (1), 84–86). [In Russian]. Some factors influencing the durability of steel-pouring moulds at the Magnitogorsk Metallurgical Combine are discussed on the basis of works records. Six types of mould used for rimming and killed steels are considered. It was found that rectangular-section moulds are less durable

than those of square section; that moulds used with hot tops for killed steel are less durable under otherwise similar conditions than those used for rimming steel; and that mould life is longest with those designs which give lowest wall temperature during filling. Higher wall temperatures due to closer positioning in the bogies also led to shorter life. Other factors studied included mould composition and microstructure, preparation, and cooling.—S. K.

Influence of Mould Temperature on the Properties of Structural Steel. N. M. Tuchkevich. (*Liteinoe Proizvodstvo*, 1956, (2), 11–14). [In Russian]. Some difficulties of obtaining reliable indications from test results of the mechanical properties of castings especially of low-alloy steel, produced by the lost-wax process are discussed and original and published experimental material is presented and analysed. Ceramic moulds at temperatures of 20, 300, 600, and 900° C. and sand moulds at 20° C were used. Mechanical properties of the cast steels were determined after hardening and tempering at 200° and at 600° C and after isothermal hardening. Elongation and cross-sectional area reduction were the properties most seriously affected by increasing mould temperature. Microstructures and inclusions for the various experimental conditions are compared.—S. K.

Use and Repair of a 200-tonne Steel Pouring Ladle. V. I. Morozov and M. T. Burnaev. (*Stal'*, 1957, (1), 38–42). [In Russian]. Improvements effected at Magnitogorsk in the design of pouring ladles are described which have enabled their capacity to be increased and repairs simplified and mechanized. Better control of ladle filling through split runners has also been obtained.—S. K.

Improvement of Killed Steel Ingots. A. N. Lekontsev. (*Stal'*, 1957, (6), 512–513). It is recommended that the H/D ratio of the ingot be increased to 4·3. A 1·5 t ingot deteriorates in quality when the slope of the faces is increased to 4%. Using feeder heads with a lining thickness of 4 in. a top discard of 12–13% is obtained and bevelling of the faces of the lower part limits bottom discard to 1–1½% of total volume. More work is required to determine optimum height.

A Study on the Seams of Billets. S. Obata. (*Tetsu to Hagane*, 1957, **43**, Mar., 416–418). [In Japanese]. The effects of top or bottom pouring, using a cylinder plate to prevent splashes and hot scarfing on the occurrence of seams in billets are discussed.—K. E. J.

FOUNDRY PRACTICE

Special or Conventional Moulding? P. Januszewicz. (*Przeglad Odlewnictwa*, 1956, **6**, June, 165–173). [In Polish]. The author analyses the factors responsible for dimensional deviations. He discusses in detail different methods of casting which reduce changes in dimensions, and surveys their range of application.

Tour of P. I. Castings (Altrincham), Ltd. D. H. Armitage. (*Foundry Trade J.*, 1957, **103**, Oct. 3, 399–406).

Mossend and Hamilton Foundries. (*Foundry Trade J.*, 1957, **103**, Aug. 29, 253–257). An account of the Fullwood Co. foundries, with plans and illustrations, is presented.

The Foundry Industry in Denmark. O. Hoff. (*Gjuteriet*, 1957, **47**, Aug., 146–147). A general review with statistics.

The Foundry Industry in Norway. T. Krogvig. (*Gjuteriet*, 1957, **47**, Aug., 152–156). A brief historical outline, and giving the number of foundries now working and a note on the use of olivine.

The Foundry Industry in Sweden. L. Villner. (*Gjuteriet*, 1957, **47**, Aug., 137–145). [In English]. (*Giesserei*, 1957, **44**, Oct. 10, 631–636). [In German]. An historical account with illustrations of early founding, maps, an account of output and an outline of present-day capacity both ferrous and non-ferrous is given.

The Foundry Industry in Finland. O. E. Huhtamo. (*Gjuteriet*, 1957, **47**, Aug., 148–151). A review with map and statistics.

Ohio Steel Foundry's First Fifty Years. C. Longenecker. (*Blast Furn. Steel Plant*, 1957, **45**, Sept., 1034–1035).

Technical Progress of Polish Foundries. S. Pelczarski. (*Przeglad Odlewnictwa*, 1957, **7**, Jan., 2–7). [In Polish]. A survey of technical progress in Polish foundries is given with special reference to problems of need for increased production of higher quality cast iron, special alloyed cast iron and magnesium alloy castings. Exothermic and insulating mixtures for risers, electric arc heating, shell moulding, special methods of casting such as diecasting and centrifugal casting are introduced and developed. The importance of the Foundry Research Institute is emphasized.

Arizona's Largest Foundry. W. G. Gude. (*Foundry*, 1957, **85**, Sept., 112–115).

Experiences of the Soviet Foundry Industry. F. Strek. [*Przeglad Odlewnictwa*, 1956, **6**, Oct., 301–307). [In Polish]. Based on his personal observations the author describes the organization and layout (with diagrams) of the large foundries in the U.S.S.R., mechanized and unmechanized.

Properties of Molten Metals. M. Olszewski. (*Przeglad Odlewnictwa*, 1955, Nov., 335–341). [In Polish]. It is emphasized that a good knowledge of the physical properties of molten metals is essential for the development of more scientific founding. Some physical properties of molten metals (e.g. surface tension, viscosity, fluidity and mobility) and their effects on casting are discussed, and a plan for a systematic investigation in this field is suggested. Examples of such investigations carried out in Poland, particularly on the use of electromagnetic pumps for molten metals, are described.—P. W.

Present State of Preparation of Standards for Iron and Steel Castings. S. Kobylinski. (*Przeglad Odlewnictwa*, 1956, Jan., 22–24). [In Polish]. After a brief description of the legal basis and scope of activities of the Polish Committee for Normalization (P.K.N.), the Polish equivalent of the B.S.I., an explanation is given that all Polish standards are divided into two classes, viz.: (1) standard "P.N.", finally approved which are either compulsory (non-compliance with these may lead to prosecution) or only recommended and, (2) standard "P.K.N.," comprising provisional new standards in various stages of preparation, and old ones under revision. Four tables of standards relating to iron and steel castings, testing materials, etc., based on the catalogue of Polish standards (1955) are appended.

The Use of Gases in the Foundry. A. Portevin. (*24th International Foundry Congress Papers*, 1957, 119–131). In the first part, the effects of gases on metal quality are described and in the second, the effects of the atmosphere or of gases in comparison with vacuum treatment in melting, holding, pouring and solidification are reviewed. Agitation and the importance of the contact surface are discussed and atmosphere control and the economics of inert gas utilization considered.

The Oxidation of Molten Cast Iron. N. Kayama and K. Nozaki. (*24th International Foundry Congress Papers*, 1957, (2), 559–586). The oxides in molten iron were examined chemically and by electron diffraction and microstructure and shrinkage cavity formation were observed. It was concluded that the oxide formed is usually SiO_2, which breaks up graphite structure and produces shrinkage cavities. Sound castings result by reducing silica content and by preventing its formation at low temperature, that is, by preventing low-temperature oxidation and then by raising the temperature as high as possible.

Investigation of the Desulphurization of Cast Iron with Magnesium. V. I. Lakomskii. (*Liteinoe Proizvodstvo*, 1957, (1), 9–11). [In Russian]. In the investigation described 2-kg samples of Armco iron were melted in an induction furnace with FeS to give a melt with 0·08–0·10% S. Radioactive sulphur was then added followed by Mg additions at regular time intervals. Samples of the liquid metal were taken after each addition and their radioactivities were determined. From the results, the relation between the quantity of Mg added to the metal and the quantity of S removed was found: this is shown graphically and the kinetics of the process are discussed. Further information was obtained from analyses of trapped vapours rising during the process.—S. K.

Prospects and Methods for Lowering the Weight of Castings. P. N. Bidulya. (*Liteinoe Proizvodstvo*, 1956 (11), 1–2). [In Russian]. The author complains that too little attention is given to practically applicable topics by research organizations and gives some examples of problems for solution in the foundry field. These deal mainly with the use of new materials and designs to reduce the weight of castings.—S. K.

Preventing Sticking with Large Castings. P. I. Shportenko. (*Liteinoe Proizvodstvo*, 1957, (2), 27). [In Russian]. Experience at the Novo-Kramatorsk metallurgical works in the production of large castings using various mould and core mixtures is briefly described. Good surfaces were obtained

with mixtures of 100 parts by weight of chromite ironstone, 2–3% of 1·27 sp.gr. sulphite lye and 5·5–6% water.—S. K.

Graphite-Talc Suspension as a Means of Preventing Sticking. I. D. Zelikman and A. S. Kurdin. (*Liteinoe Proizvodstvo*, 1956, (11), 28). [In Russian]. A suspension of 25% (by volume) of talc and 75% silvery graphite in sufficient water to give a density of 1·34–1·4 has been found to minimize sticking between moulds and iron castings. The suspension is sprayed on to the finished mould. The optimal casting temperature is 1270–1250° C for the above suspension.—S. K.

The Possibilities of Improved Castings by an Ultrasonic Method. J. Tabin. (*Przeglad Odlewnictwa*, 1956, **6**, Aug., 230–233). [In Polish]. The author gives a detailed survey of experiments so far conducted in this field in Russia, Germany and England and points out the advantages.

Production Castings and the Small Mechanized Foundry. D. Holt. (*Inst. Brit. Found.*, 1956, **7**, 65–80). The planning of a small foundry for pump parts is described from the standpoint of the production engineer. A standard range of flasks and a master plate system was adopted, equipment and operation are outlined.

Vibrating Conveyors. J. Chahbandour. (*Foundry*, 1957, **85**, Sept., 160–162, 165–166, 167). An illustrated account of foundry uses with consideration of drives and advantages.

Planning and Mechanization of Foundries in Russia. R. Chudzikiewicz. (*Przeglad Odlewnictwa*, 1955, Oct., 294–300). [In Polish]. Projects for two typical, partly mechanized foundries for the production of castings for heavy machine tools, forges, presses, rolling mills, turbines, etc., for a mechanized foundry for precision castings and for modernization of a small foundry for non-ferrous castings are discussed on the basis of recent Russian publications. Layouts, brief descriptions of processing departments and processes, and various numerical data (technical, organizational and statistical) are given.

Cupola of 600 mm Diameter with External Water Cooling. S. Fitta. (*Przeglad Odlewnictwa*, 1956, **6**, Nov., 374–376). [In Polish]. The author describes the method of external water cooling of a cupola of 600 mm diameter adopted by him and gives the results of 6 month's use. To improve the results the cooling was carried to a height of 800 mm and the thickness of the lining reduced to 123 mm. The temperature of the molten iron was 1380°–1400° C. Recommendations for further improvements are given.

Ways of Increasing Iron Temperature in Cupola Smelting. I. T. Garkusha. (*Liteinoe Proizvodstvo*, 1957, (2), 18–20). [In Russian]. Based on observations that cupolas at the "Krasnaya Zvezda" and Rostel'mash works, operating with open slag notches, have consistently produced iron at unusually high temperatures, the author has developed a cupola in which part of the combustion gases move downwards, leaving through the fore-hearth. The distribution of gas and charge temperatures up the cupola for ordinary and for the new operating conditions are considered. In experiments the effects of varying the proportion of gases flowing downwards were studied, and it was shown that the procedure was advantageous in the early but not in the later stages of the process. Lining wear was also studied and water cooling is recommended.—S. K.

The Improvement in Properties of Molten Iron. IV. The Utilisation of Oxygen Blowing for Cupola Molten Iron. T. Tottori and M. Nishikubo. (*Tetsu to Hagane*, 1957, **43**, Jan., 12–17). [In Japanese]. Repeated tests on cupola iron containing 3·7–3·8% C showed that, the more rapidly the temperature was raised by O_2 blowing and the higher the final temperature the more easily decarburization proceeded. Additions of Fe–Mn, Fe–Si or Ca–Si after blowing produced irons of high strength and heat-resistance, compared with those not so blown. An iron inoculated with Ca–Si after blowing had a tensile strength of 38 Kg/mm².—K. E. J.

Method of Controlling a Cupola-Charging Crane. V. P. Mikhailov. (*Liteĭnoe Proizvodstvo*, 1956, (6), 11–12). [In Russian]. A device and circuits are given for a crane control system which ensures that impacts of the charging mechanism against the side of the cupola are avoided. A safety device is incorporated in the system.—S. K.

Use of High-Strength Cast-Iron Castings. B. A. Efimov and V. T. Svishchenko. (*Liteĭnoe Proizvodstvo*, 1956, (4), 27–28). [In Russian]. It has been shown that with adequate care in the design of the mould system, the removal of sulphur-magnesium slags and other factors, castings of inoculated iron can be prepared which can replace steel. High-phosphorus content can be compensated for by reducing the carbon content. Details of the inoculation procedure adopted are given.—S. K.

Heat Resisting Cast Irons for Casting Furnace Accessories. A. F. Silaev and A. A. Aĭriyants. (*Liteĭnoe Proizvodstvo*, 1956, (2), 28–29). [In Russian]. The compositions and properties of some alloy cast irons for high temperature service in furnace structures are discussed. A silicon iron alloyed with chromium is recommended and directions for its production are given.—S. K.

Naturally Alloyed Khalilovo Cast Iron in Machine Construction. D. P. Glukhov, G. M. Kogan and B. M. Demkov. (*Liteĭnoe Proizvodstvo*, 1956, (4), 8–9). [In Russian]. The use of a cast iron containing 2·7–3·0% Cr and 0·8–1·0% Ni for casting cylinder blocks is described. Permissible ranges of alloying elements for various conditions are given.—S. K.

The Suitability of Indian Pig Irons for the Production of Nodular Cast Irons. P. K. Gupte, M. N. Parthasarathi, and B. R. Nijhawan. (*Trans. Indian Inst. Met.*, 1954–55, **5**, 129–158; discussion, 159–160). The influence of phosphorus on the mechanical properties of nodular cast iron was studied in the range 0·02–1% P. For high ductility in the as-cast and annealed conditions, the P contents should not exceed 0·08 and 0·2% respectively but for certain compositions 0·35% P is permissible. Methods of reducing the P content of Indian pig irons are discussed.

Mechanism of the Liquid Inoculation Process. M. M. Turbovskiĭ. (*Liteĭnoe Proizvodstvo*, 1956, (2), 15–16). [In Russian]. Analytical data obtained over a period of years for white, grey, and inoculated cast iron and experimental evidence on the mechanism of the liquid-method inoculation process are presented. Mean compositions are tabulated and frequency curves for the various elements are shown. Letting the liquid metal stand after the inoculation was shown to produce an evening-out of composition, tending to produce de-inoculation, and it is recommended that for a large casting the two solutions should be mixed during the mould-filling process. Solid and liquid inoculation processes are compared.

Form and Structure of Graphite in Cast Iron Treated with Magnesium. K. I. Vashchenko, N. A. Golovan' and R. P. Todorov. (*Liteĭnoe Proizvodstvo*, 1956, (3), 19–24). [In Russian]. Conclusions on the shape, growth process and structure of graphite in Mg-inoculated iron are drawn from a critical review of the literature and of the authors' experimental results. These latter include those obtained by the electron microscopical study of the structure of the graphite/metal-matrix boundary, which revealed magnesia cubes in the graphite itself, in the boundary zone and in the matrix. The general conclusion is that for the normal Mg content (0·04–0·05%) the graphite inclusions are spheroidal in form and sectorial in structure, most of the Mg being in the form of oxides and sulphides. Magnetic, spectroscopic, X-ray and other methods are recommended for further studies in this field.—S. K.

Treatment of Cast Iron with Magnesium in a Pear-Shape Ladle. S. I. Ryabukhov. (*Liteĭnoe Proizvodstvo*, 1956, (3), 26). [In Russian]. Experiments are briefly described in which the influence of various factors on the inoculation of cast iron with Mg were studied with the aid of a special 100-kg ladle. This was roughly the shape of a converter and was filled in the horizontal position. On tilting the ladle the iron found its way *via* a slit to a small chamber containing the magnesium. 1600 sq. mm was found to be the optimal cross-sectional area for the slit, a larger value leading to splashing and a lower value to quick erosion of the slit, although accompanied by more efficient inoculation.—S. K.

Role of the Resistance of the Medium in the Graphitization of Cast Iron. A. F. Landa. (*Liteĭnoe Proizvodstvo*, 1956, (6), 21–23). [In Russian]. After a critical consideration of some existing views on rate-controlling factors in the graphitization of iron the author's own views and their experimental bases are set out. These experiments, with specimens of identical composition and initial structure annealed under various conditions for various durations and also decarburized, showed the importance of the resistance of the metal itself to graphitization volume changes. Thus the rate can be governed by the plasticity of the metal as well as by diffusion of carbon, growth of graphite and solution of carbides.

The Possibility of Active Dephosphorization and Nitrogen Removal in the Conversion of Cast Iron to Nodular Graphite

Iron. J. Czikel. (*Giessereitechn.*, 1957, **3**, Mar., 52–54). Active dephosphorization with the elements of the first 3 groups of the Periodic System will only be achieved when it is possible to produce complex phosphides of low solubility in the iron melt which are capable of coagulation, so that they can be removed from the alloy by gravity separation.—L. J. L.

Result of an Attempt to Chill Nodular Graphite Iron. H. Herbrich. (*Giessereitechn.*, 1957, **3**, Mar. 55–56). There is no technical reason why nodular iron should not be chilled. But there would appear to be little use for the resultant castings which cannot be worked and will not withstand dynamic stress. Thermal treatment, lasting only a few hours, to produce ferritic or pearlitic structure will, however, render chill-cast nodular iron suitable even for highly-stressed machine parts.—L. J. L.

On the Quality-change of Molten Iron during Holding under Low Pressure. II. R. Yoda. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Sept., 521–524). [In Japanese]. Cast irons of varying degrees of oxidation were held under low pressure at 1200° and 1400° C, and their solidification behaviour examined. As the degree of oxidation increases, the advent of the deoxidized structure is more delayed. In irons pre-deoxidized by vacuum melting, Mg additions easily complete spheroidization. Vacuum melting is very effective for refining molten iron.—K. E. J.

Influence of Sulphur on the Crystallisation of Cast Iron. A. A. Gorshkov, S. S. Nosyreva and R. A. Sidorenko. (*Liteĭnoe Proizvodstvo*, 1956, (4), 23–24). [In Russian]. An account is given of an investigation in which sulphur distribution in cast iron before and after treatment with Mg was studied with the aid of ^{35}S. From the liquid iron (3·68% C, 2·76% Si, 0·58% Mn, 0·11% P, and 0·037% S after introduction of the isotope) the magnesium-treated metal was prepared by adding 16% of a magnesium ferrosilicon alloy, to give a metal with 2·93% C, 4·96% Si, 0·58% Mn, 0·098% P, and 0·007% S. The microstuctures were studied after slow cooling. The results showed that Mg treatment of iron of normal composition does not lead to elimination of sulphide inclusions from grain boundaries since the inclusions are already absent; attempts made to explain spheroidization by such an effect are thus baseless.—S. K.

Improving the Mechanical Properties of Cast Steel. B. B. Guljajew. (*Foundry Trade J.*, 1957, **103**, July 4, 5–10, 12). A translation from *Hochwertiger Stahlguss*. A review of defects is given with their effects on mechanical properties and methods for their prevention. Alloy additions, heat-treating, deoxidation and modifiers, rate of solidification, location of feeder heads and "pipe" are dealt with.

Technological Construction of Heat Resisting Castings. P. N. Bidulya, V. P. Desnitskiĭ and S. P. Nestertsev. (*Liteĭnoe Proizvodstvo*, 1956, (4), 1–4). [In Russian]. In this article problems in the design of heat-resisting castings of high-alloy types of austenitic steel are considered. The importance of allowing for the relatively high viscosity of such steels is indicated and the defects caused by ignoring this factor are considered. Points made are exemplified by the high pressure cylinder of a gas turbine.—S. K.

The Properties of Silica with Reference to Foundry Sands. J. J. Marais. (*Engineer Foundryman*, 1957, **21**, Feb., 50–55; discussion 55–57). The influence of the allotropy of the silica and particle size on the properties of moulding sands are discussed. The use of moulding materials other than siliceous and developments in sand practice are reviewed.

Properties of Foundry Clays. (*Canad. Mines Tech. Surveys, Phys. Metal. Division (July-Dec.*, 1956), 1957, Jan. 15, 1). A laboratory comparison of 4 types is noted.

Full Automation of Central Loam Preparation Systems. P. N. Aksenov. (*Liteĭnoe Proizvodstvo*, 1956, (6), 12–13). [In Russian]. The eventual complete automation of a foundry loam preparation system, consisting of a spent-loam conveyor, magnetic pulley, screen, elevators, bunker with feeder, mixer and bunkers is discussed.—S. K.

Proper Gating for Uniform Distribution of Mold Metal. D. Coccione. (*Western Metals*, 1957, **15**, May, 60–61). The casting of hydraulic press cylinders is discussed.

S. G. Iron—A Versatile Engineering Metal. W. W. Braidwood. (*Australasian Eng.*, 1957, **49**, May 7, 41–48). A review of production and properties, including high-temperature and thermal shock resistance, wear and corrosion resistance, and notes on uses are given.

Injection Technique in Iron Castings. (*Fonderia*, 1957, **6**, May, 205–208). [In Italian]. The technique whereby solid pulverized materials are injected into the molten metal by means of a tube is described. It is shown that by injecting CaC_2 it is possible to get an iron containing less than 0·015% S. The process can be used for castings ranging from 1 lb to 12 t.—M. D. J. B.

Inoculation of Grey Cast Iron. Relative Effectiveness of Some Silicon Alloys and Active Metals as Ladle Additions. N. C. McClure, A. U. Khan, D. D. McGrady and H. L. Womochel. (*Amer. Found. Soc. Preprint*, 1957, No. 57–2, pp. 10). The theories advanced to explain the mechanism whereby inoculation of cast irons by silicon alloys affect graphite size, mechanical properties, and chilling tendency are summarized and briefly discussed. Experiments carried out to determine the relative effectiveness of ferro-silicon, Ca-Si, Si-Mn-Zr, Si-Mn, Si, Ca and Al, are then described. Silicon was not effective as an inoculant and neither was Si-Mn free from Ca and Al. The effect of commercial ferro-silicon increased with the Ca and Al content of the metal, low Ca grades having little effect. The greatest improvement in mechanical properties was obtained with a 30% Ca-Si alloy. The next most effective inoculant was metallic Ca, added to the ladle, which improved the mechanical properties and graphite distribution, and reduced the chilling tendency and cell size. Additions of Ca also lowered the carbon content of the iron by forming a carbide with a low solubility. It is suggested that CaC_2 is involved in the formation of type A graphite. Al additions lowered the chilling tendency but did not promote type A graphite formation.—B. C. W.

Influence of the Carbon-Silicon Ratio on Graphitization of Malleable Iron. A. Taub and E. Avivi. (*Foundry*, 1957, **85**, July, 101–103). The method of production of malleable iron is explained, and experiments on the effect of variation of the alloying elements on graphitization time are described. It was found that a functional relationship exists between the C : Si ratio and the time required for decomposition of cementite (t), provided that the Mn : S relationship is satisfactory:— $C/Si = 0·5 + 0·016\ t + 0·012\ t^2$.—D. L. C. P.

Comparison of Properties of Liquid- and Air-Quenched Pearlitic Malleable Irons. Part II—Duplex Cold-Melt Process and Alloyed Pearlitic Malleable Irons. (*Amer. Found. Soc. Preprint*, 1957, No. 57–59, pp. 9). The melting and heat treatment practice, and the results obtained at eight foundries producing the various ASTM grades of pearlitic malleable iron are described and compared. Particular attention is given to comparing the effects of oil- and air-quenching.

Effects of Charge Materials and Melting Conditions on Properties of Malleable Iron. E. H. Belter and R. W. Heine. (*Amer. Found. Soc. Preprint*, 1957, No. 57–48, pp. 6). Results are given of tests on 15 heats of malleable iron designed to investigate, (*a*) the effect of pig iron variations in the charge on casting properties, mechanical properties and annealability, (*b*) the effect of variations in melting technique on the recovery of residual elements, (*c*) the correlation of silicon oxidation with changes in the recovery of residual elements, and (*d*) the effect of holding time at 2800° F on mechanical properties.

The Physical and Chemical Factors of Chill Casting. Z. Gál. (*Kohászati Lapok*, 1953, **8**, Dec., 252–254). The author reviews the influence of the dimensions of chill, especially of the wall thickness, and of chemical composition and inoculation on the properties of chilled cast iron. Reference was made to the importance of 0·3–0·4% P content to achieve a high tensile strength of the case.—P. K.

Improved Metals for Castings. (*Steel*, 1957, **140**, June 3, 98–102). Advances revealed at the First Engineered Castings Show are reported. The effectiveness of Si alloys and active metals in the inoculation of grey iron was explained. A new weldable steel with good resistance to brittle fracture was reported. Additions of P and Si to nodular and weldable iron increase the susceptibility to temper embrittlement. The cause and controlling factor of galvanizing embrittlement of malleable iron are found in its chemical composition.

Austenitic Manganese Steel Technology in an Australian Foundry. H. Thomas. (*Amer. Found. Soc. Preprint*, 1957, No. 57–36, pp. 5.) The techniques employed in the foundry of Industrial Steels Ltd., Sydney, Aust., for the production of austenitic manganese steel castings are described. Among the aspects covered are metal melting, heat treatment, chemical control and metallography.—B. C. W.

Casting Steel Parts with Closed Shrinkage Heads. D. E. Beletskii. (*Stal'*, 1957, (2), 188). [In Russian]. The author considers that the method of casting under gas pressure with closed shrinkage heads is only suitable for mass-production casting.—S. K.

Method of Calculating the Yield of Steel Castings. I. Gera. (*Kohászati Lapok*, 1954, **9**, Dec., Ontöde, 270–275). Mathematical formulæ and diagrams are given for the calculation of the yield of castings in the steel foundry.—P. K.

How to Design for Sound Stainless Steel Castings. J. L. Lessman. (*Foundry*, 1957, **85**, June, 114–117). Many diagrams are given illustrating design features which lead to sound stainless steel castings.—D. L. C. P.

Casting of Non-Ferrous Alloys and Stainless Steels (*Machinery Lloyd*, 1957, **29**, July 13, 70–71). An account of processes at Phosphor Bronze Co. Birmingham.

Good Design Cuts Casting Costs. J. W. Murphy. (*Iron Age*, 1957, **179**, June 6, 100–101). Advice on the design of stainless and high-alloy castings is given; a table lists common causes of casting flaws.—D. L. C. P.

A Suggested Re-Classification of Foundry Sands. (*Fonderia*, 1957, **6**, May, 223–224). [In Italian]. The importance of good foundry sands and consistency is stressed. Methods of re-generating sands are briefly described as well as means of storing these.—M. D. J. B.

Swedish Foundry Prepares Sand Automatically. C. G. Soderlund. (*Foundry*, 1957, **85**, June, 118–121). A new sand preparation plant built to accomplish fully automatic operation at one of Sweden's leading grey iron foundries is described. Moisture content of the sand is adjusted on the basis of dielectric properties of the used sand, measured between condenser plates.—D. L. C. P.

How to Determine Moisture Requirements of Moulding Sands. R. W. Heine, E. H. King, and J. S. Schumacher. (*Amer. Found. Soc. Preprint*, 1957, No. 57–8, pp. 5). The water requirements for obtaining maximum green strength in a moulding sand are quantitatively related to the ingredients and the sieve analysis of the sand mixture. Worked examples are given to show the operation of the method. The authors point out that for other reasons foundry sands are usually operated with a higher moisture content than that needed for maximum green strength.—B. C. W.

Properties of Moulding Sands Under Conditions of Gradient Heating. N. C. Howells, R. E. Morey, and H. F. Bishop. (*Amer. Found. Soc. Preprint*, 1957, No. 57–29, pp. 7). In order to test sand specimens under thermal conditions similar to those existing in a mould, a new hot-strength test was developed in which an induction heated pedestal was used to produce a steep thermal gradient. In this test the compressive strength of the sand decreased with increasing pedestal/sand interface temperature. Results are given for sands containing different additives to show the change of compressive strength and modulus of elasticity with heating time under gradient heating conditions with interface temperatures of 2000° F and 2400° F. Tests with a casting prone to hot tearing showed that neither hot strength nor modulus of elasticity of the sand predicted the hot tearing susceptibility of a casting. However a general correlation was observed between hot tearing and dry density of the core. Sands containing ammonium nitrate gave particularly good results since this additive prevented excessive compaction of the green sand, but on heating caused the sand to become mushy and flowable.—B. C. W.

Moving Foundry Materials. Pneumatic Conveyors: Pipeline for Sand Handling. G. A. Mott. (*Mod. Castings*, 1957, **31**, June, 49–51).

Gating of Grey Iron Castings. J. F. Wallace and E. B. Evans. (*Amer. Found. Soc. Preprint*, 1957, No. 57–9, pp. 9). The functions of an ideal gating system are enumerated. The procedure for determining the optimum pouring time is then considered in detail and recommendations for the design of gating components and procedures are made, with particular reference to the selection of the choke area. Some examples illustrating the derivation of gating systems for simple plate and bar castings are given in an Appendix.—B. C. W.

Theory of Gating System Design for Metal Moulds. J. Vaculik. (*Slévárenství*, 1957, **5**, (4), 112–113). [In Czech]. General principles of gating design, particularly applicable to gravity casting into metal moulds, are derived on the basis of physical considerations.—P. F.

A New Method for Studying Riser Requirements of Castings. J. Varga. (*Foundry*, 1957, **85**, July, 106–108). A new method is described for presenting in graphical form, the results of research on the size of risers required for castings of simple shapes. By plotting the data on log-log co-ordinates, and by selecting suitable parameters for the co-ordinates the data from several investigations were found to produce straight lines. This method enables a reduction in the number of test castings to be made when determining the size of riser needed on a particular casting.—D. L. C. P.

Some Questions On Up-to-Date Pressurizers in the Steel Foundry. J. Ferenczi. (*Kohászati Lapok*, 1956, **11**, May, Ontöde, 108–111). The author describes the Hungarian experiments on feeding of steel and iron castings at greater than atmospheric pressures; afterwards he discusses the further implications of this method.—P. K.

Increasing Yield by Heating Feeder Heads. N. P. Zhetvin, A. A. Lebed'kov, V. P. Tunkov, and A. P. Zaitseva. (*Stal'*, 1957, (7), 587–592). A method is described for intense heating of the hot tops of large ingots and heavy castings by combustion of 75% ferrosilicon in oxygen giving a 4–15% saving of metal. Exothermic mixtures for 1 t ingots are given and cropping is reduced by 2·3% with killed steel.

Metal Penetration Into the Mould Material. Z. Nagy. (*Kohászati Lapok*, 1955, **10**, Jan., 7–12). Reference is made to a mathematical correlation between the pressure of the molten metal column and the surface stress of the molten metal penetrating the capillary channels of the mould. Metal penetration may be promoted by the following factors; high metal column, thick metal layer, high casting temperature, unfavourable cutting of the runner, unsuitable placing of the hot tops, improper refractories, coarse grain mould material and unfavourable casting design.—P. K.

The Clamping of Chill Moulds. (*Fonderie*, 1957, May, 225–228). The various types of clamps that can be used with chill moulds are illustrated and described in detail.

Air Setting Speeds Coremaking. D. R. Chester. (*Steel*, 1957, **141**, July 22, 120–123). The technique and advantages of the air-setting or cold-set process of coremaking are described. The process uses an oil binder which develops green strength in the sand mixture when it oxidizes with air. It is a suitable process for jobbing foundries making large and medium size castings; less rodding is required and considerable timesaving results.—D. L. C. P.

"Compo" and Chamotte Moulding. R. Wright. (*E.S.C. News*, 1957, **11**, Summer, 5–9). The composition, properties, production, moulding technique and comparison of compo with bonded sands are reviewed in an illustrated article.

Practical Comparison of Case Mould Materials. M. K. Young, J. N. Trant and R. Hamilton. (*Amer. Ceram. Soc. Bull.*, 1957, **36**, July, 260–261). Gypsum, elastomers, phenolic, polyester and epoxy resins are briefly reviewed.

Automatic Spray Boosts Core Mould Output. (*Iron Age*, 1957, **180**, July 11, 102–103). The operation and advantages of an automatic core blowing and spraying installation in a cast iron pipe foundry are described. Two men can produce over 200 6-in. cores every hour, and automatic spraying gives a better coating than hand spraying.—D. L. C. P.

The Effect of Coal Dust on the Properties of Moulding Mixtures and the Surface Quality of Castings. J. Dlezek. (*Slévárenství*, 1957, **5**, (4), Appendix). [In Czech].—P. F.

The Carbon-Dioxide Process. G. E. Parramore. (*Brit. Found.*, 1957, **50**, July, 359–371). A review of the process and testing methods used by I.C.I. Billingham maintenance foundry and of developments elsewhere is given. The use of local sand and sand recovery methods are described with notes on a 5-t liquid CO_2 tank installation.

An Evaluation of the Carbon-Dioxide Process. J. E. O. Little. (*Brit. Found.*, 1957, **50**, July, 372–379). A discussion of the advantages and disadvantages and economic aspects of the process compared with oil-sand and dry-sand moulding is given. Special cases where the process is of particular value are noted.

What You Should Understand About The Carbon Dioxide Process. W. E. Gruver jun. (*Foundry*, 1957, **85**, June, 106–110). The nature of the process is explained and the effects of the following variables considered:— amount and properties of sodium silicate, gassing time, water present, fine material, hygroscopic materials, sand grain size. Such understanding of the process is essential for efficient practice.

Development of the CO_2 Process. P. Williamo. (*Fonderia Ital.*, 1957, **6**, June, 240–242). [In Italian]. The author describes the CO_2 process for the preparation of moulds and cores and shows the rapid progress which the process has made quite recently. The advantages are briefly reviewed.

Increased Productivity in Steel Foundries Resulting from the Use of Rapidly Drying Moulding Mixtures Containing Water-Glass. J. Fiala. (*Slévárenství*, 1957, **5**, (4), 104–111). [In Czech]. The preparation of rapidly drying moulding mixtures, their properties and uses in core and mould making are described. Casting technology is discussed using concrete examples, and some types of faults in castings are considered. Some of the mixtures are dried by conventional methods, but dry much more quickly than the sand mixtures commonly used; others are hardened by means of CO_2.—F. F.

Mould Hardness: What it Means. R. W. Heine, E. H. King, and J. S. Schumacher. (*Amer. Found. Soc. Preprint*, 1957, No. 57–20, pp. 5.) The average mould hardness and the green strength of a sand are shown to be closely related. For clay-saturated or fully bonded sands a general relationship exists which is independent of the type of sand, sieve analysis, moisture level and sand additions. For clay-poor sands however the relation depends on clay content and the factors influencing clay distribution. The reason for this behaviour is briefly discussed.—B. C. W.

Ethyl Silicate Moulds. A. Torry. (*Iron Steel*, 1957, **30**, Aug., 414–418). An account of Darwins Ltd. "Truprocess" moulding methods for the production of die-casting dies by the Shaw process is given in some detail with illustrations of finished dies and of various stages in the operations.

The New Look in Cleaning Rooms. C. King. (*Mod. Castings*, 1957, **31**, June, 52–54). Improvements at the Dodge cylinder block dept. are described.

Foundry Improves Cleaning. (*Steel*, 1957, **141**, July 8, 114). An installation for moving castings from the grinding station to the cleaning mills and on to the inspection station is described. It is operated by one man.—D. L. C. P.

Blast Cleaning Makes Fast Work of Large Castings. (*Iron Age*, 1957, **179**, June 27, 101). A blast chamber which can clean 100 tons of castings at a time is described. Eight wheels bombard the work with 3200 lb of steel shot per min.

An Original Method of Increasing Production. H. Haynes. (*Brit. Found.*, 1957, **50**, July, 379–382). A description of the top-pouring method for the production of liner castings is given.

Production of Big and Complicated Grey Iron Castings. B. Jagrik. (*Kohászati Lapok*, 1956, **11**, (8), 177–181). The author discusses raw materials and moulding materials, as well as the moulding and casting technology used for the production of large and complicated grey iron castings in Hungary.—P. K.

Special Problems in Heavy Steel Castings Production. Z. Nagy. (*Kohászati Lapok*, 1955, **10**, May, Ontöde, 107–112; June, 131–134). The author deals comprehensively with moulding materials as well as with the shrinkage and stress formation in heavy steel castings.—P. K.

Guide to Engineered Castings. (*Mod. Castings*, 1957, **31**, June, 55–70). Design recommendations and tables of mechanical properties of alloys are given. Notes on special service conditions (heat and corrosion), uses of various metals, casting processes and dimensional characteristics are discussed and tabulated. (66 references).

Production of Rail Soleplates for Reinforced Concrete Sleepers. G. P. Efimov. (*Stal'*, 1957, (7), 627–628).

Four 8½ Ton Cast Steel Ladles. (*Commonwealth Eng.*, 1957, **44**, Apr. 1, 85–86). The process of casting is described.

Production of Machine-Tool Castings. G. W. Nicholls. (*Brit. Found.*, 1957, **50**, June, 296–308). Standardization and control of all materials and operations are coordinated for the production of heavy castings and the investigation and prevention of defects is carried out. Many special cases are reviewed.

The Manufacture of Aircraft Engine Castings. C. W. Hicks. (*Brit. Found.*, 1957, **50**, Aug., 408–418). A careful account of the production of cast-iron support rings and nozzle boxes as well as Al and Mg parts is given, with an account of inspection and testing.

The Production of Ferchromite Castings. K. Weber. (*Slévárenství*, 1957, **5**, (4), 97–103). [In Czech]. The structure, properties, production and uses of heat resisting high-Cr castings and the influence of the content of Cr and other elements on the properties are discussed. The production of ferchromite, casting and moulding techniques are considered.—F. F.

Experiences in the Production of Centrifugal Cast Cylinder Liners. T. Budinszky and J. Gerédi. (*Kohászati Lapok*, 1954, **9**, May, 97–102; June, 121–130). The authors discuss the technology of moulding and centrifugal casting of cylinder liners in Hungary and review these in other countries. They comment on the importance of sufficiently high melting, tapping and casting temperatures as well as of 500–600° C chill temperature. To ensure the former, melting in electric furnaces rather than in cupola furnaces is suggested.—P. K.

The "Alfin" Process of Centrifugal Casting. (*Fonderia*, 1957, **6**, May, 217). [In Italian]. A brief account is given of a process which enables two metals to be cast, into a mould which is then rotated. This gives an "inner" and an "outer" layer to the casting which may have many technical advantages.—M. D. J. B.

Shell Moulding. T. Budinszky. (*Kohászati Lapok*, 1953, **8**, July, Ontöde, 145–149). A short review is given of the development of shell moulding techniques, materials and equipments. The accuracy of the metal mould machining must be 0·08–0·05 mm. The results of shell moulding experiments for steel casting are not yet satisfactory.—P. K.

Mechanisation for Shell Moulding. K. H. Coombs. (*Engineer Foundryman*, 1957, **22**, June, 51–57). A review of the process as used in South Africa. Pattern equipment, moulding materials, resin binders, stripping agents, mould closing, running, feeding, and pouring, pre-coated sands, tolerances and defects are discussed. Other topics were brought up in discussion.

Epoxy Compound Saves Foundry Work. (*Steel*, 1957, **140**, June 17, 132). Uses of an epoxy resin compound containing Al, "Metalset A2," are described. It is used to salvage defective castings, cement gates to patterns and make modifications in patterns.—D. L. C. P.

Temperature Drop in Pouring Ladles, Part II. V. Paschkis and J. W. Hlinka. (*Amer. Found. Soc. Preprint*, 1957, No. 57–41, pp. 5). The temperature drop in pouring ladles containing liquid iron or steel has been calculated for various conditions assuming as a limiting case that the metal is completely mixed so that a uniform temperature drop occurs. The results are given in the form of curves relating holding time to pouring temperature and preheating temperature. The problem of estimating holding times in practice is briefly considered.—B. C. W.

Provisions Against Burning of Mould Sand on the Castings. A. Toth. (*Kohászati Lapok*, 1953, **8**, Dec., 254–257). Burning of the mould sand on the castings can be reduced by the following four factors: (1) High heat resistance and (2) fine grain size of the sand. (3) Low pressure and (4) low oxide content of the molten metal. Quartz increases, metal oxides, lime, limestone, magnesium oxide and pyrites decrease the heat resistance of the mould sand. In view of this, more than 1% Fe_2O_3, MgO or CaO and more than 1·5% MgO + CaO are harmful. The burnt sand is amorphous after rapid cooling but crystalline after slow cooling. The former is easier to remove from the casting than the latter. The mould dressing is either a gas generator, containing much carbon, or heat-resistant with high quartz or ZrO_2 content. The former is used for thin, the latter for thick castings.—P. K.

New Achievements in Shell Moulding Technology. W. Kowalski. (*Przeglad Odlewnictwa*, 1956, **6**, Nov., 329–334). [In Polish]. The author states the advantages and disadvantages of shell moulding and gives the basic principles of making shell moulds. He discusses the difficulties of core making and surveys the different moulding sands and their costs. He gives a detailed description of an automatic machine for shell mould production and discusses methods of joining. A detailed description of the Lintzburg Foundry Co. of U.S.A. is given.

Coal Silt in Shell Moulding and Coremaking Sand. A. Potocki and J. Bielatowicz. (*Przeglad Odlewnictwa*, 1956, **6**, Nov., 345–346). [In Polish]. The authors give an account of an experimental attempt to replace coal dust and clay substance by more readily available and cheaper coal silt which was delivered direct from the coal mines. The results were very satisfactory. The observations lasted a few months and were carried out on casts from a few kg to 4·5 t. The exact ratio of sand to coal silt and the humidity are determined by the purpose for which the mixture is used.

The Mechanization of Shell Moulding. R. Chudzikiewicz. (*Przeglad Odlewnictwa*, 1957, **7**, Feb., 35–39). The author gives a survey of different methods of shell moulding, classifying them into 4 groups. He also gives F. Gaty's classification of shell moulding machines and describes the principles of operating the moulding machines of Sutter type S.P. 1000, S.P. 1100, F 32/45 and two-layer shell type F 50/65. Details of methods of using adhesive in the shells made by the Polygram Casting Company are given.

New Method for Casting Thin-Walled Large Parts. N. N. Rubtsov and E. S. Stebakov. (*Liteinoe Proizvodstvo*, 1956, (12), 8–10). [In Russian]. A new method of casting is described for producing large, thin-walled objects at present made from sheet and sections. The method consists essentially of bringing two halves of the mould, arranged vertically, and containing a small excess of metal, rapidly towards each other so that the excess metal is squirted out, the movement is stopped to give the required wall thickness. In practice the two halves are pivoted at the bottom. The rapid movement of metal gives a clean, fine-grained structure.

Analytical Solution of the Simplest Problems of the Solidification of Castings of Different Configuration. N. G. Girshovich and Yu. A. Nakhendzi. (*Liteinoe Proizvodstvo*, 1956, (12), 13–18). [In Russian]. This is the concluding part of an analytical study of the solidification of castings. Solidification under conditions of ideal contact with superheating and the existence of a crystallization interval, and solidification when a heat-insulating layer (gap) is present are considered. A nomogram is given for determining the period of solidification.—S. K.

New Approach to the Analysis and Definition of Casting Defects. J. Pribyl. (*Przeglad Odlewnictwa*, 1957, **7**, Jan., 7–12). [In Polish]. Definitions of casting defects and the conditions for their formation are discussed. The critical period, the speed and time of formation of defects are described, and suggestions of eliminating them are given.

Defects of Machine Components made from Spheroidal Graphite Cast Iron, and Means to Avoid Them. L. Sofronie. (*Met. Constr. Masini*, 1956, **8**, (9), 43–48). [In Rumanian]. The characteristic defects of these castings are analysed and classified. A few of these defects are reviewed in this article and remedies proposed for their elimination.—L. H.

Hot Cracks and Tears in Malleable Iron Castings. R. W. Heine and F. W. Jacobs. (*Foundry*, 1957, **85**, Sept., 120–125). Five general classes of variables are discussed which may produce tears or cracks in conjunction with normal contraction. All are within the control of the foundryman except faults in design. Unfed internal and external shrinkage; mould tears and cracks, over-size or swelled or undersize; Core tears, design and stress tears and miscellaneous types are considered. Miscellaneous causes include gating and risering faults, poor equipment, and metal temperature and condition.

Investigations of Moulding Material Defects on Castings. W. Heinrichs and F. Völpel. (*Techn. Mitt. Krupp.*, 1957, **15**, Sept., 95–98). Various casting defects due to the penetration of metal into the mould material were examined under the microscope; the photomicrographs are reproduced and discussed.—L. D. H.

Prevention of Hot Cracking in Steel Castings with the Aid of Shrinkage Ribs. G. Kniaginin. (*Przeglad Odlewnictwa*, 1956, Feb., 33–39). [In Polish]. Causes and conditions under which hot cracking occurs in castings are discussed. Guiding principles and numerical data for the application of shrinkage ribs, worked out on the basis of theoretical considerations and experimental results, are given.

A Comparison of Costs for Molten Iron from Induction Furnaces and Cupolas. M. Rydinger and B. Lundborg. (*Gjuteriet*, 1957, **47**, Aug., 173–179). [In English]. High frequency and mains frequency furnace data from several foundries are evaluated and hot blast and cold blast cupolas of similar capacity are compared. The data suggest trends rather than give actual results as the calculations are somewhat uncertain.

Application of Statistical Methods in the Establishment of Standards for Size Deviations of Castings. R. Krzeszewski. (*Przeglad Odlewnictwa*, 1955, Nov., 341–347). [In Polish]. Results of a statistical analysis of dimensional deviations in iron castings, which revealed considerable discrepancies between these deviations and those recommended as a guide by the Polish standards authority, are reported and discussed. An opinion is expressed that the discrepancies found are due to the mistaken application of the law of similarity, in the standards for castings of all shapes and sizes while this law is valid for geometrically similar cylindrical objects only. It is suggested that statistical methods should be used in the establishment of new standards.

Ventilation in Foundry Shops. P. V. Ogorodnikov. (*Liteinoe Proizvodstvo*, 1956, (11), 9–11). [In Russian]. Some measures for keeping working conditions satisfactory when production is increased with the same floor space are outlined. Ventilation, the use of heat screens, concrete plates for ventilators and dust-settling chambers are briefly discussed.—S. K.

Health Hazards in Foundries and their Prevention. A. Ahlmark and H. Öhman. (*24th International Foundry Congress Papers*, 1957, 1–27). Air pollution, radiation and noise are surveyed. The effects of dust, fumes and gases are described and limits for these and for heat and noise are suggested. Silicosis and siderosis are reviewed with reference to foundry operations. **The Foundry Environment.** H. J. Weber. (29–43). New materials and processes are discussed from the point of view of occupational hazard, Be, synthetic resins and radioactive materials are included. **Foundry Ventilation.** K. Nilsson and Y. Bovin. (45–81). An extensive review of general and spot ventilation systems is given. **Air Pollution and Dust Control in a Foundry.** A. Piggi. (83–117). An account of the Fiat Foundry at Turin with cast iron and Al departments is given. Machines and equipment are listed and smoke and dust exhaust systems are described. Test data over a long period are given.

Stop Foundry Accidents. (*Mod. Castings*, 1957, **32**, Sept., 35–46). A general review of hazards of all sorts with recommendations for safe working.

Need for Improvement in Working Conditions in Finishing Shops of Cast Iron Foundries. J. Holtorp. (*Przeglad Odlewnictwa*, 1956, Jan., 25–26). [In Polish]. The author describes the unsatisfactory conditions of safety and hygiene in the finishing shops and discusses the causes and frequency of accidents and occupational diseases. Special emphasis is placed on dust, harmful gases and noise. Suggestions for improvements in hygienic conditions and prevention of accidents are listed.

VACUUM METALLURGY

High Vacuum Electric Annealing Furnaces of Large Size. E. Zeitz. (*Elektrowärme*, 1957, **15**, Aug., 287–289). An illustrated account is given of furnaces working up to 1350° C with steel charges up to 25 t at vacua of 10^{-3}–10^{-4} mm. A 120 kW bell-type tempering furnace holding 5 t takes about 30 h to warm up to 1300° C and 35 h to cool to 500° C with a power consumption of 500 kWh/t. A 300-kW car bottom type furnace holding 25-t of sheets requires ~ 6 days for a cycle at a consumption of 420 kWh/t at 1100° C.

A Vacuum Arc Melting Furnace for Refractory Metals. O. Z. Rylski and H. V. Kinsey. (*Arcs in Inert Atmospheres and Vacuum*, 1956 *Symposium of Electrothermics and Metallurgy Division of the Electrochemical Society, Apr.-May*, 69–76). The furnace works at reduced pressure in a protective atmosphere.

Vacuum Remelting of Superalloys and Steels by the Consumable Electrode Process. W. W. Dyrkacz, R. S. DeFries and R. K. Pitler. (*Arcs in Inert Atmospheres and Vacuum*, 1956 *Symposium of Electrothermics and Metallurgy Division of the Electrochemical Society, Apr.-May*, 97–117). Properties of remelted metals are studied. Electrodes used were air-melted alloyed cast ingots. Comparison with air-melted alloys is made for mechanical properties, gas content and inclusions and by forged upset tests.

The Importance of the Vacuum Casting of Steel in Metallurgical Practice. Z. Eminger, F. Kinsky and Z. Kleteška. (*Hutnické Listy*, 1957, **12**, (8), 755–767). The only method so far found to be satisfactory is vacuum casting followed by cooling at atmospheric pressure, holding of the reduced pressure was impracticable. Vacuum casting is useful only at pressures below 40 mm and is most suitable for large ingots. Considerable improvements can be expected at pressures below 5 mm. Shortened production time and better control and more uniform properties can be expected for large castings, though a considerable improvement of properties over the values obtained by current methods is not to be expected.

Vacuum Treatment of Electric Steel. G. M. Borodulin. (*Metallurg*, 1957, (5), 16–18). Two installations at Dnyeprospetstal are described, trials have been carried out both in the ladle and while pouring from one ladle to another, the second being in a vacuum chamber. In 1956, 26,000 t were vacuum treated, the greatest improvement being with transformer steel. H_2 was reduced by 2–44%, still more being removed in the two-ladle process. In ball bearing steel 78% of the H_2 was removed.—R. S.

Titanium and Aircraft-Quality Steels. W. Jessop and Sons, Ltd. (*Overseas Eng.*, 1957, **31**, Oct., 86). Production by vacuum-melting is outlined. Compositions and properties are not included.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Diversification in Commercial Heat Treating. F. J. Rizzo. (*Indust. Heating*, 1957, **24**, Aug., 1672, 1675, 1677, 1679, 1681, 1682, 1690). An account of H_2 annealing, carbon restoration and carburizing, bright hardening of stainless steels and interrupted quenching as carried out by the L.R. Heat Treating Co., N.J.

Modernization of Wheel Heating Facilities Johnstown Plant, Bethlehem Steel Co. F. R. Pullen. (*Iron Steel Eng.*, 1957, **34**, July, 94–99). Details are given of the new heating furnaces for the wheel mills at Johnstown. The mill operation is described.—M. D. J. B.

Temperature Control of Laboratory Heat-Treatment Furnaces. A. P. Ryl'nikov. (*Zavodskaya Laboratoriya*, 1956, **22**, (8), 999–1000). [In Russian]. A temperature-control system for laboratory muffle furnaces which does not require electronic or mechanical potentiometers of certain types is described. The system has been successfully used for four years on heat-treatment furnaces at the Central Research Institute for Ferrous Metallurgy.—S. K.

Control of Atmosphere Generators. R. R. Swain. (*Iron Steel Eng.*, 1957, **34**, Sept., 152–159). Measurement, analysis, sampling and control systems are described with reference to exothermic generators.

Infra-red Analysers Monitor Furnace Atmospheres. J. L. Garrison. (*Iron Steel Eng.*, 1957, **34**, July, 145–146). An industrial infra-red analyser for the measurement of gas components in a mixture is described briefly.—M. D. J. B.

Electronic Heating and Automation. M. T. Elvy. (*J. Brit. Inst. Radio Eng.*, 1957, **17**, Aug., 443–462). A general review of radiofrequency apparatus and methods including metallurgical applications.

Heat Treatment of Pearlitic Malleable Cast Iron. M. Dubowicki, W. Sakwa, and S. Pieprznik. (*Przeglad Odlewnictwa*, 1956, **6**, Sept., 258–267). [In Polish]. Investigations of how the properties of pearlitic malleable cast iron are influenced by temperature and time of heating before hardening as well as by tempering, were conducted in Czestochowa Polytechnic both under laboratory and industrial conditions. Some of the conclusions are: (1) 850° C is considered the most suitable temperature for oil hardening; (2) by tempering the hardened pearlitic malleable cast iron an increase in hardness of 120–200% was obtained compared with the initial state; (3) the greatest tensile strength was achieved by tempering at 650°. Results are tabulated and curves shown.

Influence of the Original Structure and Composition on the Properties of Cast Iron after Heat Treatment. I. F. Kurtov and N. V. Sadin. (*Liteinoe Proizvodstvo*, 1956, (11), 20–23). [In Russian]. Experiments on a total of six types of cast iron are described. The cast specimens were isothermally hardened under identical conditions and then subjected to mechanical and wear-resistance tests and microstructures and residual austenite were determined. The irons were grey, malleable and Mg-treated, originally in the ferritic and pearlitic states. Curves showing the influence of heat treatment temperature on the elongation, tensile and compressive strength, wear-resistance and residual austenite are shown and discussed.—S. K.

The Effect of Heat Treatment on the Structure of Low-carbon Steel. S. Ono. (*Tetsu to Hagane*, 1957, **43**, Mar., 361–362). [In Japanese].—K. E. J.

Improving the Strength of Low-Carbon Steel by Thermal Treatment. N. V. Shmidt, Z. N. Krasil'shchikov, N. T. Pavlenko and E. N. Shvach. (*Stal'*, 1957, (9), 833–837). МСт3 steel can be brought up to the mechanical properties of MK steel in the untreated condition. It contains 0·15% C, 0·49% Mn, 0·23% Si, 0·042% S, 0·017% P, 0·13% Cr, 0·07% Ni and 0·21% Cu. Structures are shown and uses indicated.

Gas-Air Flame Hardening Increases Gear Strength. J. P. Bates and C. A. Turner, jun. (*Indust. Heating*, 1957, **24**, Aug., 1526–1528, 1530, 1532, 1534). Gears for tractor towing winches are flame-hardened by the process described. Hardness surveys of tooth surfaces are shown.

The Hardening of Steels 40 C 10 and 40 CN 12 with Special Reference to the Thermal Treatment of Drilling Rods. D. Rascev, N. Antonescu and M. Comeaga. (*Met. Constr. Masini*, 1956, **8**, (12), 21–29). [In Rumanian]. The compositions of the two steels 40CN12 and 40C10 are given. The results of the investigation show that substitution of steel 40C10 in many cases for 40CN12 is of advantage.—L. H.

Studies on Gas Carburisation. On the Relation between Toughness and Depth of the Carburised Layer. N. Kamuro. (*Tetsu to Hagane*, 1957, **43**, Mar., 232–233). [In Japanese] For SH85B and SAE 4320 steels, the variation in case depth with carburizing time at 900° C is shown. Corresponding values are given for the variations in Charpy impact value and hardness, for the cases of carburizing and quenching at 800° C, with and without intervening furnace cooling.—K. E. J.

Hard and Very Hard Monophase Surfaces Obtained by Diffusion. H. Buckle and R. Moissan. (*Galvano*, 1957, Jan., 27–32). The authors discuss a solid/solid diffusion method for surface hardening. The systems Cu/Zn Cu/Sn and Mo/C are discussed in this connection. Results of microhardness tests on the surfaces are given.—R. P.

Thickness of the Work-Hardened Layer in Shop Peening. M. Ya. Shashin. (*Vestnik Mashinostroeniya*, 1956, (11), 47–50). [In Russian]. The article is a review of previous work done in this field, especially of the method of measuring the depth of the layer by the electric and magnetic properties of the metal. The apparatus used in the investigation was the EMKP for X-ray analysis. A full description of the procedure is given and graphs and statistical data listed.—L. H.

Strengthening by Boriding the Working Surface of Low Alloy Steels. V. D. Taran and L. P. Skugorova. (*Vestnik Mashinostroenniya*, 1957, (5), 62–65). Experiments on three marks of steel 30 XGSA, 12XN2A and 40 X, are described. Results show that the most suitable temperature for the operation is between 920 and 950° and its duration 6 h. This assures the highest resistance of the surface. Under the same conditions steel 12 X N2 A gives the best results.

Peculiarities of the Dissociation of Austenite in High-Frequency Hardening and Subsequent Tempering. I. N. Kidin. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (3), 530–536). Contrary to the impression that low temperature ageing at 150° removes the advantage obtained by high frequency hardening and leads to a different structural arrangement from that obtained by the usual hardening, the authors have proved that the condition of the hardened material is nearer to the tempering condition immediately after high frequency hardening than after the usual hardening. Hence to obtain the optimum combination of resistance and viscosity the heating must be adjusted to the process of high frequency hardening and not applied as in the ordinary process.—L. H.

The Heat Treatment of High Speed Steel Tools. I. Burcioiu. (*Met. Constr. Masini*, 1955, (12), **7**, 41–45). To standardize production, ten groups of high-speed steels have been established and the heat treatment of these steels is described. Use is made of the Russian process of heating with high frequency currents, which, when properly applied, are said to insure a much quicker dissolution of the carbide. Repeated reheating with repeated cold treatment is said to double the strength of these tools.—L. H.

Induction Harden Small Lots for Extra Benefits. R. T. Hook. (*Iron Age*, 1957, **180**, 126–127). Local hardening techniques are outlined and equipment, changes in materials used and savings are noted.

High Frequency Heating for Surface-Hardening, Local Annealing and Brazing. F. L. Gladwin. (*Electrowärme*, 1957, **15**, Aug., 290–294). Generators are described made by Wild-Barfield with many advantages and a guaranteed life of the valves in excess of 6000 h.

Theoretical Principles Governing the Electrical Production of Heat. A. Leitner. (*Maschinenbau Wärmewirtschaft*, 1957, **12**, Sept., 245–261). Theoretical considerations governing

the electrical production of heat are discussed, and typical applications described. The laws of heat transfer are treated in the second half of the paper, and a bibliography is appended.

The High Intensity Arc: Electrode Requirements for Metallurgical Application. S. Korman and C. Sheer. (*Arcs in Inert Atmospheres and Vacuum, 1956 Symposium of Electrothermics and Metallurgy Division of the Electrochemical Society, Apr.-May*, 184–188). Electrodes consist of mixtures of C and non-conducting ore and the ratio is subject to conflicting requirements. Vaporization rates vary linearly with current density, slope depending on latent and specific heats. Balance of factors is necessary to obtain the best erosion resistance and yield.

Africa's First Nitriding Plant. (*Engineer Foundryman*, 1957, **22**, Aug., 35). A brief account of equipment and operation is given.

Effect of [Niobium] and Nitrogen on LCN155 Heat-resisting Steel. On Precipitation in Tempering after Solution Treatment. T. Masumoto. (*Tetsu to Hagane*, 1957, **43**, 251–253). [In Japanese]. Increases in hardness at 1000° C and after solution-treatment, and in max. hardness produced by additions of Nb and N_2 are given. The grain size number is increased only by Nb. The types of precipitation after solution-treatment at 1150° C are considered for the steel without additions, with additions of Nb or N_2 and with addition of both.—K. E. J.

FORGING, STAMPING, DRAWING, AND PRESSING

Damage to Forging Dies During the Forging Process. W. Wroblewski. (*Hutnik*, 1956, **23**, Dec., 447–450). [In Polish]. The author discusses various types of damage to forging dies such as: breaking, abrasion of surfaces, softening of the tool-steel at higher temperatures, cracking of the working surfaces, distortion and adhesion of the forged metal to the surface of the dies.

Faults in Free Hot-Forged Pieces and Their Causes. G. Isac. (*Met. Constr. Masini*, 1957, **9**, (5), 9–12). [In Rumanian]. These defects fall into two classes; (*a*) defects due to heating and cooling, and, (*b*) faults in forging operations. Among the first category are mentioned unsuitable heating, overheating, burning, decarbonizing, unsuitable cooling and flaking.—L. H.

Progress in Recent Times in the Technology of the Hot Working of Steels, and What Should Our Attitude Be Towards This Development. V. M. Zoltan. (*Met. Constr. Masini*, 1957, **8**, 77–86). [In Rumanian]. The suggestions made refer mainly to the process of manufacture, and also to the utilization of Ti, which is recognized as the future metal for the light transport industry.—L. H.

The Closing of Internal Cavities in Forgings by Upsetting. A. Tomlinson and J. D. Stringer. (*J.I.S.I.*, 1958, Mar., **188**, 209–217). [This issue].

Directional Properties caused by Forging and Deformation Ratio due to Heat-Treatment with Different Heavy Sections of Tool Steel. H. Honma. (*Tetsu to Hagane*, 1957, **43**, Mar., 231–232). [In Japanese].—K. E. J.

The Fundamentals and Application of Form and Thread Rolling. C. T. Appleton. (*Steel Processing*, 1957, **43**, Aug., 440–446; Sept., 498–503, 525). A cold forging technique for producing threaded forms is described. Tools and equipment, flat die, planetary and cylindrical die machines are described. Thread rolling on screw machines, dies and rolls, and preferred forms are also outlined. The preparation of blanks is then reviewed and a considerable section on applications of the process concludes the series.

Precision in Instantaneous Countersinking in Sulphur Cast Iron Mouldings. T. F. Terlikova. (*Testnik Mashinostroenniya*, 1956, (11), 38–41). [In Russian]. Countersinking in mouldings of sulphur cast iron, in one operation, as a finishing method has been widely applied, especially for parts where openings of 4th degree precision sufficed. A previous investigation has shown that various factors could contribute to increasing precision. In the main these are the deformation of the elastic system of the machine-tool complex, wear and tear, the temperature of deformation of the various tools, etc., errors in setting, conditions of work, etc. The article deals with investigations to establish by mathematical and statistical methods the dependence of the precision of work on the conditions under which the operations are conducted and on the geometry of the tools.—L. H.

Defects in Pressed Steel Components. C. Gazdaru. (*Met. Constr. Masini*, 1956, **8**, (5), 16–21). An analysis of defects shows that the great majority of them are due to defects in the raw materials used and about one-third only are due to faults in the final manufacturing processes.—L. H.

Laws Governing the Drawing of Metals in the Presence of Lubricants. S. Ya. Veiler and V. I. Likhtman. (*Doklady Akad. Nauk. S.S.S.R.*, 1957, **114**, (6), 1224–1227). [In Russian]. Copper and brass strips were drawn between freely rotating cylindrical rolls lubricated by oleic acid and other lubricants. The relation between mechanical constants and the properties of lubricants are outlined.—S. I. T.

Some Test Results on the Redrawing of Cold-Rolled Sheets. T. Akamatsu. (*Tetsu to Hagane*, 1957, **43**, Mar., 235–236). [In Japanese]. Investigations are reported into the effects of speed on drawability of cold-rolled sheets, using both round and flat punches, and the relationships between the original blank diameter and the maximum load in redrawing.

On the Properties of Phosphate Coatings for Cold-Drawing. C. Misumi. (*Tetsu to Hagane*, 1957, **43**, Mar., 366–368). [In Japanese]. The influences of solution temperature (up to 100° C) and time of reaction (up to 24 min) on the weight of phosphate coating and the loss in weight of treated iron are shown graphically.—K. E. J.

Studies on Cold Extrusion of Steel. V. Experimental Results of Cold Backward Extrusion in 3000-ton Hydraulic Press. S. Tsujimoto. (*Tetsu to Hagane*, 1957, **43**, Mar., 238–239). [In Japanese]. The variations of extruding power and ram stroke with extrusion time are shown. Mechanical properties are tabulated for two specimens from the extrusion, one from the bottom and one from the wall, for the cases of extrusion normal and parallel to the fibre structure.—K. E. J.

Extrusion of Profiles With Non-Uniform Section. I. Petrenco. (*Met. Constr. Masini*, 1956, **8**, (8), 53–56). The article gives the history of how an industrial plant, which has been given the task of extruding a complicated (Al alloy) profile, has overcome the theoretical and technical deficiencies in its production. The theoretical, physical and mathematical investigations, as is generally the case, are of a very complete nature.—L. H.

ROLLING-MILL PRACTICE

An Experimental Study on Yield Stress of the Theoretical Formula for Rolling Load. K. Kato. (*Tetsu to Hagane*, 1957, **43**, Mar., 383–385). [In Japanese].—K. E. J.

Trends in Lubrication Design of Rolling Mills. S. L. Norton. (*Davy-United Eng.*, 1957, **1**, Sept., 20–39). An illustrated review.

Use of Reserve Frictional Forces in Rolling on a Blooming Mill. B. P. Bakhtinov. (*Stal'*, 1957, (9), 805–809). Increased reduction can be obtained by correcting for frictional forces the relation between angle of bite at the beginning and during a fixed rolling schedule.

Selection and Economy of Equipment for Blooming and Slabbing Mills. K. Schlesinger. (*Iron Steel Eng.*, 1957, **34**, July, 63–74). The author describes 3 typical mills with outputs of 200,000, 500,000 and 10^6 t/year and shows how capital investment per ton/year of product drops as capacity increases. Typical layouts are discussed.—M. D. J. B.

Flow Production of Structural Steel. (*Brit. Eng.*, 1957, **40**, Oct., 124–127). An illustrated account of Boulton & Paul Ltd. works at Norwich and the improvements installed since 1953.

Automation in the Production of Rail Joints. C. E. Chapman. (*Iron Steel Eng.*, 1957, **34**, Aug., 81–86). The automatic preparation of rail ends is described and details are given of the plant and equipment used.—M. D. J. B.

Flexible High Speed Bar Mill at Phoenix. H. H. Campbell. (*Iron Steel Eng.*, 1957, **34**, July, 77–81). The paper describes the 12 in. modern bar mill at The Phoenix Manufacturing Co. The mill is designed to roll a wide range of products.

On the Rolling of Rod and Bar by a Continuous Rolling Mill Separately Driven. T. Harada. (*Tetsu to Hagane*, 1957, **43**, Mar., 385–388). [In Japanese]. Data on roll wear, power consumption, rolling speeds and torque, cross-sectional area of the rolled material, and relationships between them, are given for vertical and horizontal stands.—K. E. J.

American Steel and Wire's New Rod Mill. W. V. Magee. (*Iron Steel Eng.*, 1957, **34**, Sept., 160–165). The mill at Cleveland, Ohio, rolling rod up to 1½ in. dia. is described and produces > 100-t/h.

Revaluation of the 84-in. Plate Mill Modernized at Alan Wood. F. J. Bampton. (*Iron Steel Eng.*, 1957, **34**, Sept., 113–116). Experience of running the modernized mill for 2 years is recounted. Each operation is discussed throughout the process reporting changes made and results obtained.

Intensification of the Process of Rolling Plate Blooms Based on Investigations Conducted on a Blooming Mill in the Lenin Metallurgical Works in Cracow. W. Leskiewicz. (*Hutnik*, 1956, **23**, Nov., 401–403). [In Polish]. The dependence of the horizontal flow of rolled metal on the distance between the rolls was experimentally investigated. A new rapid method of determining the profile of the rolled bloom is described. A vessel filled with refractory clay (of plasticine consistency) is pressed against the side of the bloom. Analysis of profiles thus obtained at various stages of the rolling process led to the elimination of unnecessary operations and reduction of the number of stages. A new method of calibration of the mill was developed in which the number of stages was reduced, the side walls of the bloom were vertical and rolling time was cut down by 20%.

On the Manufacture and Working of Hot Rolled Stainless-Steel-Clad Sheet Steel. M. J. Bottema. (*Lastechn.*, 1957, **23**, July, 125–9). [In Dutch]. The article concerns the manufacture of sheet slab clad with stainless by hot rolling, the adhesion of the layers and the methods of welding. The problem of heat transfer between the layers is dwelt upon and there are some comments on the desirable economic thickness of such clad materials.—F. R. H.

Effect of Hot- and Cold-Rolling on the Austenite Grain Size of Case-Hardening Steel. H. Ishida. (*Tetsu to Hagane*, 1957, **43**, Mar., 241–242). [In Japanese].—K. E. J.

Surface Defects on Sheet from Killed Steel Ingots. K. F. Lüdemann and H. Löscher. (*Abhandlungen der Deutschen Akademie der Wissenschaften zu Berlin*, 1956, 127–151). The authors show how surface defects can arise on steel sheet. The influence of melting and casting conditions on the ingots are discussed. Means of preventing these defects include efficient deoxidation and preheating of the ingots. (20 references).—R. P.

Electrical Features of a Hot Strip Reversing Rougher with Close Coupled Edger. J. H. Greiner and A. Mozina. (*Iron Steel Eng.*, 1957, **34**, July, 88–92). This article describes the electric equipment of the 4-high reversing rougher with close coupled vertical edger for the roughing train of Granite City Steel Company's 90 in. hot strip mill.—M. D. J. B.

Modern Development of the Ward-Leonard Principle and Applications in Steelworks. G. Ovens and C. A. Dodd. (*J.I.S.I.*, 1958, Mar., **188**, 266–276). [This issue].

A New Static Control System. F. C. Fennell. (*Iron Steel Eng.*, 1957, **34**, Aug., 125–131). The author describes a static control system for a reversing roughing stand of a hot strip mill. The control system programmes reversing mill operations from punched cards and is compact and labour saving.—M. D. J. B.

Pneumatic-Hydraulic Control System for Self-Aligning Coilers. A Chelyustkin. (*Stal'*, 1957, (9), 822).

Developments in the Cold Rolling of Oriented Transformer Strip up to 500 mm Wide. M. Žídek and V. Dědek. (*Hutnické Listy*, 1957, **12**, (9), 853–864). The present process of manufacture is described and conditions of cold working are analysed. Optimum intermediate annealing conditions and thermal range for working are determined. A process for rolling 2·9–3·4% Si steel to 0·32 mm thickness, annealed and cold rolled with an intermediate anneal is developed. Watt losses are $\sim$ 0·35–0·40 W/lb at B = 10,000.

Effects of Cold Rolling on the Properties of 8% Chromium Steel. V. G. D'yakov. (*Stal'*, 1957, (9), 837–840). Analyses of X8, X8T, X8BT and X8CM steels are given and cold rolling up to 50% deformation had no effect on their hardness. At 30–40%, however, hardening is especially marked, but the slope of hardness against degree of reduction dies away at higher values. In steels with comparatively great initial hardness increase is relatively less. X8 steel at 50% reduction increased in hardness by 70%, while X8CM steel increased 40%, these are maximum and minimum effects in the steels used, however the absolute hardness of X8 is always lower than that of X8CM.—R. S.

Deformation Structure in the Cold Rolling of Low Carbon Steels. K. V. Grigorov and G. P. Blokhin (*Fisika Metallov i Metallovedenie*, 1957, **4**, (1), 161–170). [In Russian]. The process of the texture formation of steels containing 0·1% carbon changes during deformation approximately proportionally to the rate of deformation. It takes place gradually. During weak deformations it grows chiefly at the expense of the relative volume of the crystals with the orientation (112) [110] but upon further increase in the deformation it grows at the expense of the relative volume of the crystals of both orientations.—L. H.

Research Work of the All-Union Scientific and Research Institute for the Tube Manufacturing Industry. Ya. E. Osada and S. I. Borisov. (*Stal'*, 1957, (9), 861–863).

The Effect of Altering the Angle of Feed on the Quality of Tubes. A. Z. Gleiberg. (*Stal'*, 1957, (9), 810–815). In diagonal tube rolling increase of feeding angle reduces the number of flaws on the internal surface and also reduces number of passes.

Studies on the Rolling of Seamless Steel Tubes. II. Measurement and Analysis of Every Load on the Plug Mill. Y. Yamada. (*Tetsu to Hagane*, 1957, **43**, Mar., 338–340). [In Japanese]. The driving system of the Aetna-type plug mill is described, and the moment of inertia of every part is calculated. A system for recording rolling force and torque is described. Results for rolling 2-in. gas pipes include rolling force, rolling torque and mandrel force for a range of rolling temperatures.—K. E. J.

Electric Drive Systems for Bar and Rod Mill Shears. A. J. White and J. A. Read. (*Iron Steel Eng.*, 1957, **34**, Sept., 166–173). Various types of shears are described and power requirements are calculated.

Recent Developments in Rolling Mill Gear Design. L. Allen. (*Davy United Eng.*, 1957, **1**, Sept., 3–19). Wear of pinions, effect of accuracy of cutting and production, trends in reduction gears, heat treatment and case hardening, couplings and expected life are reviewed.

Crowned Tooth Gear Type Couplings—Development and Application for the Steel Industry. F. W. Zurn. (*Iron Steel Eng.*, 1957, **34**, Aug., 98–113). The functions, methods of design and maintenance of crowned tooth gear flexible type couplings are examined and the advantages and uses of this type of coupling are described.—M. D. J. B.

Rolling Mill Spindle Maintenance at Gary Steel Works. I. D. Pottorf and R. H. Davis. (*Iron Steel Eng.*, 1957, **34**, Sept., 79–82, Discussion 82–84). The two types of spindles, "wobbler" and "universal" ended, and the general plan of the rolling mill stands are indicated and typical failures, by fatigue or impact, and their repair are reviewed and fractures are shown. Other points, and practice at other mills, are brought out in discussion.

Modernized Suspended Machine for Metal Trimming. V. K. Pasechnik and M. V. Suralev. (*Stal'*, 1957, (9), 816).

The Safe Operation of Bar Reeling and Straightening Machines. W. W. Hickman, G. M. Flather, E. S. Harrison, R. K. Mawson and E. L. Tinley. (*British Iron and Steel Federation Publication*, [1957], pp. 32). An illustrated review of machine design and methods of safe operation is given.

Treatment and Re-Use of Mill Scale Wastes. G. A. Pettit. (*Iron Steel Eng.*, 1957, **34**, Sept., 144–150, Discussion, 150–151). Steps taken by Armco Steel to conserve water and abate pollution of the Miami river are described. A special sedimentation basin was constructed. Ferric sulphate is used as coagulant and lime is added and sludges removed and filtered.

MACHINERY FOR IRON AND STEEL PLANT

Use of Repeating Devices for Automatic Transfer of Oval Sections. P. A. Skobolov and S. S. Smirnov. (*Stal'*, 1957, (9), 816–819).

Rapid Handling by Means of The "Steelmaster." L. Gascuel. (*PACT*, 1957, Aug., 333–334). The author gives a short description of the "Steelmaster" transporter for handing pipes, girders etc., and the uses to which it is put. The average load is 8 tons.—R. P.

Distribution System Studies in Steel Mills. D. Dalasta and F. Nolan. (*Iron Steel Eng.*, 1957, **34**, Aug., 87–95). The authors discuss electrical distribution problems in steel mills, short circuit, load and stability problems and methods of modernization and expansion. (9 references.)—M.D. J. B.

LUBRICATION

Standards and Classification of Lubricants for Steel Mills. J. D. Lykins. (*Iron Steel Eng.*, 1957, **34**, Sept., 101–104). A coding system is described.

Measurements for the Consistency of Lubricating Grease. B. Holliday. (*Iron Steel Eng.*, 1957, **34**, July, 100–104). The author discusses factors which affect the consistency of greases. Additives, water content, fillers and bodying agents, manufacture, chemical factors and the character and percentage of soap are examined and commented upon.

Improving the Quality of Asphalt Lubricant for Roll Necks in Thin Sheet Rolling Mills. B. Otta and J. Teindl. (*Hutnické Listy*, 1957, **12**, (9), 803–808). Testing is surveyed and consumption in various types of mill is given. New asphalts with 90–110 softening point, flash point above 310° C and 0·3 max. evaporation at 300° in 1 h and only traces of ash are described.

Silicone Lubricants for the Steel Industry. W. H. Ragborg and W. H. Badger. (*Iron Steel Eng.*, 1957, **34**, Sept., 105–107, Discussion 107–108). Classification of liquids and greases and their properties and uses are outlined.

WELDING AND FLAME CUTTING

Workshop Layout for Resistance Welding. V. Gardavský. (*Zváranie*, 1957, **6**, (7), 204–210). [In Czech]. Equipment, works' organization, use of weld-testing machines, layout of control sheets for welds, etc., are considered.—P. F.

Welding under Carbon Dioxide. Z. Duben. (*Zváranie*, 1957, **6**, (7), 201–203). [In Czech]. The scope and application of CO_2 as protecting atmosphere are discussed with special reference to Soviet experience and with an eye to the possibilities of the method in Czechoslovak industry.—P. F.

The Problems of Welding in CO_2 of Chromium Nickel Austenitic Steels and of Heat-Resisting Alloys. B. I. Medovar and I. N. Rublevskii, (*Avtom. Svarka*, 1957, (3), 70–84). [In Russian]. A very detailed study is reported giving ample details of tests and results, which are tabulated. Among the many obvious conclusions and suggestions which result, the following may be mentioned. On welding in CO_2 a slight oxidation takes place in the protecting atmosphere. There should therefore be a minimum content in the metal seams of Si, H_2 and P so as to ensure an increased resistance of the austenitic steels to the formation of heat cracks. Welding in CO_2 can be used in the manufacture of articles from austenite-stable steels. Nevertheless the investigations showed the need for further study.—L. H.

High-current Arc-welding Processes for Mild Steel Under Protective Atmospheres. A. Moreau. (*Rev. Soudure*, 1957, **13**, (1), 1–11). Two processes are described using A and CO_2 respectively, with an account of the reactions involved and tests on the materials.

Semi-Automatic Welding With a Horizontal Electrode. V. Hromádko. (*Zváranie*, 1957, **6**, (7), 195–201). [In Czech). Butt-welds in plates were made by inserting between them an electrode in the shape of a rail. After ignition at one end the electrode is slowly consumed leaving a slag-protected seam behind.—P. F.

Semi-automatic Welding of Thin Steel with Fusion Electrodes. I. I. Zaruba. (*Avtom. Svarka*, 1957, (3), 9–19). [In Russian]. Semi-automatic welding in practice falls much below the speed of the automatic operation. Nevertheless it can be carried out on steels between 0·8 and 2 mm thickness, with thin wire electrodes of 0·5–0·8 mm with flux and in a protecting atmosphere of CO_2. Tests carried out on different brands of steel for different welds, at different speeds, with varying electrodes, etc., supply ample data justifying this semi-automatic welding as a practical proposition.—L. H.

An Investigation of the Transfer of the Electrode Metal in the Welding Arc. I. P. Patskevich. (*Avtomaticheskaya Svarka*, 1954, (2), 33–41). [In Russian]. The fusion of the welding rod is accompanied by a marked evolution of carbon monoxide and metal vapours. These may be liberated from the molten welding rod directly through the open surface without the formation of blow-holes (especially when uncovered electrodes are used) and through the medium of blowholes in the case of thickly coated electrodes. The gas which is liberated directly through an open surface creates forces of reaction which cause the movement of a droplet from the electrode and impart a definite acceleration to it. These forces are the fundamental forces which determine the transfer of the electrode metal when welding without closing the arc circuit. The transfer of metal in the welding arc and the dimensions of the electrode droplet depend on many factors which affect the amount of CO and vapours which are formed when the welding rod melts. Such factors include the type of current, the chemical composition of the electrode, coating and flux and the length of arc. It was found that the theoretical diagrams of the development of gas bubbles and of the removal of the droplet during welding were confirmed by experimental investigation of the dimensions of the droplet and by macro-analysis of longitudinal sections of the electrode after they had partially melted. This experimental work has some practical bearing on electrode wastage.—R. S.

Reliability of a Feed System for the Electric Arc. S. M. Emdina. (*Avtomaticheskaya Svarka*, 1954, (2), 48–55). [In Russian]. A successful method of harmonic balance for the analysis of controllers having an intermittent action in welding units is described.—R. S.

New Developments Abroad in the Field of Weldable Steels. J. Němec. (*Zváranie*, 1955, **4**, (9–10), 272–276). [In Czech]. The article deals primarily with the properties, uses and limitations of Carilloy and Fortiweld.—P. F.

Conference on the Weldability of Low Carbon Steels. F. Faltus. (*Zváranie*, 1955, **4**, (4), 113–114). [In Czech]. At the Conference, held in January 1955, papers dealing with weldability, effects of impurities in submerged-arc welding, brittle fracture in welded steels, and with the safety of welds in structures were read. Some recommendations for changes in Czechoslovak Standard Specifications relating to welding are also listed.—P. F.

Testing the Weldability of Structural Steels by the 28 Method of the Welding Research Institute (Bratislava). J. Lombardini. (*Zváranie*, 1955, **4**, (6), 162–168). [In Slovak]. A detailed description of the 28-Weld Test is given and its applications are discussed. The test depends essentially on measurements of the impact strength of samples prepared from a weld in a prescribed manner, the weld having been made under specified conditions.—P. F.

13Kh3GNMA Welding Wire for the Automatic Welding of 30KhGSA (0·90% C) Steel. T. M. Slutskaya. (*Avtomaticheskaya Svarka*, 1956, (2), 12–17). [In Russian]. It was found that the best type of welding rod for the automatic welding of 30KhGSA steel (C 0·31, Si 1·11, Mn 0·94, Cr 0·87, S 0·019, P 0·025) was a rod containing C 0·11, Si 0·32, Mn 0·98, Cr 2·5, Ni 0·91, Mo 0·35, Ti 0·02. This ensured uniformity of weld strength when the steel was heat-treated to 70 t tensile.—R. S.

Modern Production Technology of Electrodes used for Weld-Surfacing. J. Vidrma. (*Zváranie*, 1956, **5**, (6), 174–178). [In Czech]. Details are given of the development of compound Cr–W–V steel electrodes in which efficient use is made of expensive alloying elements. The outer coat is alloyed.

Welding of Grey Cast Iron With the Bimetal Electrode Developed by the Czechoslovak Welding Institute. L. Lakatoš. (*Zváranie*, 1956, **5**, (6), 182–187). [In Slovak]. Details are given of the methods of use and applicability of the electrode, which has been designed primarily for welding and repairs of grey cast iron without pre-heating.—P. F.

The Welding Electrode VUS-Bimetal 55. S. Horváth. (*Zváranie*, 1956, **5**, (7), 208–213). [In Slovak]. The development and use of a Cu–Fe electrode for welding grey cast-iron without pre-heating are described. The electrode, meant for mass production, is equivalent in performance to a bronze electrode but does not necessitate the use of expensive alloying materials, e.g. tin.—P. F.

Some Experiments on the Substitution of Low Carbon Ferro-Manganese in Flux of Coated Arc Welding Electrodes by High Manganese Core Wire. D. R. Dhanbhoora, S. Visvanathan, and S. N. Anant Narayan. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 301–308). To avoid using imported ferromanganese an electrode containing 0·12% C, 0·3% Si, 3·0–3·5% Mn was tested and found to give good results with a coating containing 1–2% Fe powder.—A. D. H.

On the Suitability of E391 Type Electrodes for Welding Thick-Walled Bodies Destined for Use at High Temperatures and Pressures. J. Němec. (*Zváranie*, 1956, **5**, (10), 290–296). [In Czech]. Welding technology with the electrode containing 18% Cr, 9% Ni, 1·8% Mo, and 0·8% Nb is discussed, with particular reference to welding of stabilized 18% Cr, 9% Ni steels intended for use up to 550° C. Further data,

relating mainly to ferrite control in the welds, are given in the discussion of the article by A. Kleander.—P. F.

Welding With Electrodes Having Non-Slag-Forming Coatings. V. Hromádko. (*Zváranie*, 1956, **5**, (8–9), 236–243). [In Czech]. Following publication of the use of such electrodes in the April, 1954 issue of Schweisstechnik, similar electrodes were developed and tested in Czechoslovakia. The results of the tests are discussed, the advantages and disadvantages of the method are pointed out, and care is advised in using the electrodes if high quality, strong welds are desired. Further improvements, following new development of the method, are foreseen. The results are compared with parallel tests with conventional electrodes.—P. F.

Some Physico-Mechanical Properties of Welding Flux in the Solid State. A. I. Chvertko. (*Avtomaticheskaya Svarka*, 1956, (3), 79–87). [In Russian]. The physico-mechanical properties of flux were studied from the point of view of structure, shape, dimensions and weight of the grains and also friability and specific weight. These properties depend on the grade of flux and its granulometry. The data are used for designing fluxing apparatus.—R. S.

Aluminium-Alloyed Welding Rod Wire for Automatic Welding. B. S. Kasatkin, N. I. Kakhovskii and Yu. N. Vakhnin. (*Avtomaticheskaya Svarka*, 1955 (6), 54–63). [In Russian]. The use of Al-alloyed welding rod for the automatic welding of O.H. mild steel and low alloy steel improves the mechanical properties of welds, especially resistance to cold shortness and the condition subsequent to welding and mechanical ageing. The recommended composition of welding rod is C 0·10 max, Mn 0·4 to 0·6, Si 0·1 max, Al 0·4 to 0·6, S 0·03 max. and P 0·03 max. Al-alloyed electrodes also promote arc stability, thus reducing stresses in the weld and porosity. Al in the wire promotes the transfer from flux to weld of Mn and Si especially when arc voltage is increased.—R. S.

The Segregation of Some Alloying Elements in Welds. A. A. Rossoshinskii. (*Avtomaticheskaya Svarka*, 1957, (1), 37–39). [In Russian]. Data are presented relating to the direction of segregation of some alloying elements in welds. The relationship between the direction of segregation and the phase diagrams of these elements alloyed with iron is shown.

The Effect of Hydrogen and Oxygen on the Formation of Pores When Welding Stable Austenitic Steels and Nickel. B. I. Medovar and Yu. V. Latash. (*Avtomaticheskaya Svarka*, 1957, (1), 14–30). [In Russian]. The steels examined contained approximately 23–24% Cr, 18–23% Ni, with and without additions of 2·5–3% Cu and Mo. It was found that O_2 does not cause porosity during the welding of purely austenitic steels but that it does give rise to porosity in the welding of Ni due to the reaction of nickel oxide with N_2 and also because of the oxidation of carbon. H_2 causes pores in purely austenitic welds of stable austenitic steels. In the A arc welding of nickel, however, with a W electrode, H_2 prevents porosity. The addition of TiO_2 to the flux containing CaF_2 eliminates porosity due to H_2 in purely austenitic welds. The H_2 is transformed to HF by TiF_4. Powerful oxidizers are useful for reducing the H_2 contents of austenitic welds, such oxidizers being added to the flux.—R. S.

The Effect of Carbon on the Dendritic Heterogeneity of the Distribution of Sulphur in Welded Joints. L. A. Poznyak. (*Avtomaticheskaya Svarka*, 1957, (1), 3–7). [In Russian]. Radiographic and other methods of investigation have shown that an increase in the carbon content in welded joints increases the dendritic heterogeneity of the distribution of S. With $> 0{\cdot}20\%$ C in the weld, sulphides occur which are not encountered at low carbon content.—R. S.

Testing the Weldability of Age-Hardening Low-Alloy Steels. K. Mazanec and K. Pawera. (*Zváranie*, 1956, **5**, (6), 166–167). [In Czech]. A new cracking-test developed for use with welded stock thicker than 1½ in. is described. This was found suitable by correlating with results obtained in practice.

Cladding by Means of Cast-Iron Welding Rods. R. Rohan. (*Zváranie*, 1956, **5**, (8–9), 263–267). [In Czech]. The technology of welding with cast iron welding rods, particularly in repairing worn or otherwise damaged castings, is discussed.

Some Methods of Following Welding Processes in the Weld Zone in Automatic Submerged-Arc Welding. A. Havalda. (*Zváranie*, 1955, **4**, (8), 226–231). [In Slovak]. A number of methods amenable to the study of temperature distributions in the weld zone are discussed, e.g. by means of colours and chalks which undergo colour changes, thermocouples connected to scanning devices, etc. The scopes and limitations of the various methods are reviewed.—P. F.

The Stability of Weld Metal in Relation to Hot Cracking During Automatic Welding. S. A. Kurkin. (*Avtomaticheskaya Svarka*, 1954, (2), 23–32). [In Russian]. The weld metal can be rendered sufficiently resistant to the formation of hot cracks by careful control of its chemical composition and this can be effected both by the development of new types of fluxes and by new methods of welding which make it possible to control, within wide limits, the proportion of base metal and weld metal in the joint. To determine the most suitable material and the optimum welding conditions, the resistance of the weld metal to hot cracking must be established under various conditions of welding. A fairly reliable method of determining this characteristic is described. It entails a comparison of the increase in deformation associated with failure in the brittle range and the deformation of the weld metal under given conditions of welding.—R. S.

Control of Weld Quality. F. Wiesner. (*Zváranie*, 1956, **5**, (8–9), 250–253). [In Czech.] Magnetic, ultrasonic, X-ray and other methods of testing are discussed and their respective advantages and scopes are outlined.—P. F.

Time-Temperature Effect on Properties of Weld Heat-Affected Zone in Type 347 Stainless Steel. E. F. Nippes, B. Schaaf, W. L. Fleischmann, and R. L. Mehan. (*Welding J.*, 1957, **36**, June, 265s–270s). The heating and cooling cycles occurring in metal-arc welding of 1½-in. plate were established and the potentially troublesome zone was located at about 2400° F. Samples were tested in simulated cycles to 2450° F and the effects on grain size and carbide phase determined.

Control of Submerged Arc-Welding by Means of Magnetic Amplifiers. P. M. Zuffa. (*Zváranie*, 1956, **5**, (7), 212–213). [In Slovak]. Based on a report in *Elektrotechnik*, No. 1, 1956, the article describes how deviations from the set value of the current passing through the electrode can be utilized for the automatic control of the uniformity of the weld.—P. F.

Radiation Methods for Checking Welds. F. Wiesner. (*Zváranie*, 1956, **5**, (10), 306–309). [In Czech]. The use of X-rays, gamma-active isotopes and high-temperature radiography for checking weld-quality are discussed.—P. F.

Method of Testing Welded Joints of the Reinforcement of Reinforced Concrete Structures. A. Ya. Brodskii. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1220–1224). [In Russian]. Procedures for testing welded joints including complicated ones, in reinforcing bar structures for tearing, shear and twisting loads are described.—S. K.

Static Tensile Strength of Spot Welded Joints. A. N. Dorofeev. (*Avtomaticheskaya Svarka*, 1956, (2), 58–67). [In Russian]. It was shown that any repeated stressing greatly increases the strain-hardening effect. There is a non-uniform distribution of stresses among the various spot welds, causing overstressing in some parts of the entire spot welded joint. Fatigue stressing of the welded joint will cause failure more rapidly than static loading.—R. S.

Static Tensile Strength of Longitudinal Butt and Lap Welded Joints. B. F. Lebedev. (*Avtomaticheskaya Svarka*, 1956, (2), 48–57). [In Russian]. Tests showed the advantages of two-way butt welding instead of lap welding for the horizontal joints of vessels. Lap welded joints, even of high grade material, sometimes failed because of low ductility.—R. S.

Approximate Determination of the Zone of Active Stresses During Welding. I. P. Trochun. (*Avtomaticheskaya Svarka*, 1954, (2), 71–75). [In Russian]. Two formulæ are put forward for the approximate calculation of the width of the zone of active stresses during welding of two strips. The calculations obtained by means of the formulæ are in good agreement with the experimental data.—R. S.

Relative Behaviour of Notch-Toughness Tests for Welded Steel. W. J. Murphy, W. D. McMullen, and R. D. Stout. (*Welding J.*, 1957, **36**, June, 307s–312s). Notched slow-bend, NRL drop-weight and V-notch Charpy tests were carried out on 11 structural steels and the appearance of the fractures was also examined. Correlation between the drop-weight NDT and Kinzil 1% lateral contraction temperatures was found and a plot of NDT against Charpy 10 ft-lb temperature was also made and interpreted.

The Effect of Hydrogen in the Weld Metal in Welding Age-Hardening Low-Alloy Steels. K. Mazanec. (*Zváranie*, 1955, **4**, (9–10), 264–269). [In Czech]. Results of experiments and theoretical interpretations of these are given. A

definite enhancement of the tendency to crack formation was found if hydrogen was present in the Ni–Mo–V steel electrodes. Precautions necessary in welding and in the manufacture of electrodes to minimize contamination of the welds with hydrogen are outlined.—P. F.

Some Points Relating to Stresses in Hard Welded-on Surfaces. V. Linhart and K. Löbl. (*Zváranie*, 1956, **5**, (6), 163–166). [In Czech]. A study of weld surfaces used in high-pressure steam generators in power plants in Czechoslovakia is reported. Welded-on surfaces applied to an austenitic base show no reduction of internal stresses in comparison with similar surfaces applied without an intermediate austenitic layer. Surfaces made from the alloy Real 096 require furnace-annealing; torch-annealing followed by cooling in sand is inadequate.—F. F.

Effect of Carbon on Abrasion-Resistance of Weld Metal. E. I. Leinachuk. (*Avtomaticheskaya Svárka*, 1956, (3), 58–64). [In Russian]. The effect of carbon on the hardness and abrasion-resistance of weld metal containing 0·17–2·54 C, 0·39–0·54 Mn, 0·6–0·15 Si, 0·009–0·018 S and 0·032–0·045 P was examined. When the carbon content was increased from 0·17 to 2·21, abrasion resistance was increased more than four times. For a given hardness, the higher the carbon content, the greater the resistance to abrasion which also increases as the sorbitic structure changes to troostite and free cementite is precipitated, increasing in quantity.

Brittle Fracture of Welded Vessels and Other Constructions. G. V. Raevskii. (*Avtomaticheskaya Svarka*, 1955, (6), 3–18). [In Russian]. Until vessels are filled, all their welded joints exhibit residual stresses caused by welding. These stresses are equal to the yield point of the steel and coincide, in direction, with the casting stresses. The fact that most vessels do not fail, however, shows that high residual stresses, as a rule, do not affect the strength of a welded construction of mild steel. Susceptibility to cold shortness and to the thermal effects of welding can be reduced during steel making. Stress concentrations caused by overlapping joints, non-continuous welds, manholes, etc. can be eliminated by good design. Cold shortness caused by operations under onerous conditions of climate can be avoided by gently heating the vessels. Steel quality is of prime importance and the Russians consider that the failure of Belgian welded bridges, for instance, is due to the use of low grade basic Bessemer steel. The Soviet Petroleum Ministry has specified killed steels instead of rimming steels for welded vessels.—R. S.

Brittle Fracture of Welded Metal Constructions. V. V. Shevernitskii and G. V. Zhemchuzhnikov. (*Avtomaticheskaya Svarka*, 1955, (6), 19–29). [In Russian]. The authors examine the effect of three factors on the static tensile strength of welds (notches, residual stresses, design). Although notches may reduce the static tensile strength, they do not reduce it below the specified limits for temperatures down to − 60° C. Residual stresses must be present to cause further reduction in tensile strength. Brittle fracture can be largely avoided by good design.—R. S.

Formation of Porosity in Automatically Welded Joints. L. S. Sapiro. (*Avtomaticheskaya Svarka*, 1954, (2), 78–82). [In Russian]. Experiments are described with a view to studying the mechanism of the formation of porosity in welded joints. The formation of the gaseous phase in the molten weld metal is possible only on the boundary between the liquid and the solid phase. There are no foundations for the assumption that pores are formed in the welded joint only by way of gas bubbles which appear during solidification. The beginning of a gas channel originates from the boundary of the liquid and solid metal at which the bubble emerging from the gas channel is nucleated and maintained by flotation forces and the gas channel itself is caused by active diffusion of the gas from the liquid metal to the source bubble.—R. S.

Strength of Silver Brazed Joints in Mild Steel. W. G. Moffatt and J. Wulff. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 2, 442–445). The authors have studied the dependence of the strength of silver brazed joints in mild steel on the geometry of the joint. In cylindrical specimens with transverse joints, ultimate tensile strength is not merely a linear function of joint thickness, but also of the thickness/diameter ratio.—G. F.

Guide to Better High-Temperature Brazing. W. Feduska. (*Iron Age*, 1957, **179**, May 16, 116–118). Choice of brazing alloy, heating rate and protective atmosphere for fluxless high temperature brazing applications are dealt with.

Oxygen Cutting of Steel in Continuous-Casting Installations. O. Sh. Spektor. (*Stal'*, 1956, (11), 1042–1048). [In Russian]. An account is given of the laboratory and full-scale investigation of the cutting process of carbon and alloy steels and of the application of the results in the development of equipment for use with continuous casting. The horizontal cutting of hot steel was found to follow the same laws as ordinary oxygen cutting. A process and equipment have been developed for horizontal cutting (i.e. suitable for vertical continuous-casting installations) of the steels at speeds sufficient for full-scale casting. Further work is required before the system could be applied to continuously-cast high-alloy steels. Relationships and nomograms are given for calculating the main process conditions for cutting hot billets with a depth of cut of from 100 to 300 mm.—S. K.

Automatic Welding of Low Alloy Steels of High Resistance in a Medium of CO_2. B. S. Kasatkin, N. I. Kakhovskii and Yu. N. Bakhnin. (*Avtom. Svarka*, 1957 (3), 39–43). For welding in a CO_2 atmosphere of low alloy steels NL 2, CXL4 and 14XTC, Si–Mn wire of type 10 G. 2 CA is recommended; the use of this wire ensures metallic seams with equivalent properties as specified by the recognized specifications.—L. H.

Automatic Welding of Thin Steel in Carbon Dioxide. I. I. Zaruba and A. G. Potap'evskii. (*Avtom. Svarka*, 1957, (3), 22–27). [In Russian]. In practice, automatic electric arc welding of thin metal makes great use of fluxes and the protecting media of inert gases, especially in argon arc-welding. The tests described in the article demonstrate that thin low-carbon steel (0·8–2 mm) can be perfectly satisfactorily welded in a protecting atmosphere of CO_2 with thin wire CB 10 GC. Also thin chrome steels with not less than 18·5% Cr can also be welded with thin wires CB 1X18H9T and the welds possess good mechanical properties and are resistant to intragranular corrosion.—L. H.

Automatic Welding of Heat-resisting Steel 15XMA in CO_2 Atmosphere. B. S. Kasatkin, N. I. Kakhovskii and Yu. N. Vakhnin. (*Avtom. Svarka*, 1957, (3), 28–38). [In Russian]. Electric arc welding in CO_2 of heat resisting steel 15XMA can be successfully carried out. The technological and mechanical properties of the welded metals are very good within a temperature range between 20–550°, and the seams maintain stable structure and properties during ageing in the range 500–550°. The use of this welding process increases productivity in comparison with automatic welding with fluxes. Also electric welding in a gaseous atmosphere includes less slag in the seams, thus giving a greater resistance of the seams at working temperatures.—L. H.

Inert-Gas Tungsten-Arc Welding SAE 4130 Steel Sheet.—I. C. A. Terry and W. T. Tyler. (*Welding Metal Fab.*, 1957, **25**, Oct., 397–402). Experiments leading to a satisfactory technique are reported with a discussion of factors leading to porosity or cracking. Further work on weld cracking is needed.

Stainless Steel Welding. (*Torch*, 1957, Summer, 9–11). General notes with illustrations are given.

On the Gas Weldability of Rimmed Strip Steel. T. Fujioka. (*Tetsu to Hagane*, 1957, **43**, Mar., 368–370). [In Japanese].

Effect of Some Alloying Elements on the Weldability of Low Alloy Structural Steel. H. Tamura. (*Tetsu to Hagane*, 1957, **43**, Mar., 304–306). [In Japanese].—K. E. J.

On the Weldability of Low-carbon Steel Sheets in the Unionmelt Welding Process. K. Kakutani. (*Tetsu to Hagane*, 1957, **43**, Mar., 260–262). [In Japanese].—K. E. J.

Gas Burners with Air Suction. P. Bazil and F. Traian. (*Met. Constr. Masini*, 1956, **8**, (6), 19–25). Two types have been constructed and are here described; the smaller with a capacity of 45m³/h and the larger of 100m³/h for use in repair shops of railways and other industrial concerns.—L. H.

Pipe Manufacture and Use. P. D. Thomas. (*Iron Steel Eng.*, 1957, **34**, July, 84–87). Developments in the manufacture and use of pipes are discussed. Butt and lap-welding techniques and seamless pipe production are described and the prospects of future markets are reviewed.—M. D. J. B.

Production of Electrically Welded Pipes from Cut Rimming Steel Strip. Ya. A. Mednikov and E. M. Krichevskii. (*Stal'*, 1957, (9), 819–822).

Wire for Welding Wear-resisting Steels in a Medium of CO_2. I. K. Pokhodnya. (*Avtom. Svarka*, 1957, (3), 51–54). [In Russian]. The composition of electrode rod for welding steels P18, 3X2B8 and X12BF has been worked out. The

welding of high-carbon wear resisting steels proceeds conveniently with the aid of a powder wire electrode. The loss of electrode material from coal gas and from spray can be minimized by introducing some Ti into the composition of the powder wire. Any positive influence of Si is very small.

Oxidation of the Components in Austenitic Electrode Wire on Welding in a CO_2 Atmosphere. B. I. Medovar, D. A. Dudko and I. N. Rublevskii. (*Avtom. Svarka*, 1957, (3), 64–69). On welding in CO_2 with austenitic electrode wire containing 18–25% Cr, oxidation of these elements practically does not occur, and this welding process is therefore preferable to manual welding with the open arc. Si and Mn in small proportions are oxidized, the Si less than the Mn, but the proportions oxidized vary according to the concentration of these elements in the alloys. Ti is oxidized to nearly 50% in concentrations of up to 1% but in concentrations of up to 3% its oxidation does not exceed 30%.—L. H.

On the Evaluation of the Weldability of Austenitic Nickel–Chromium Steels Stabilized with Niobium for Use at High Temperatures. J. Němec. (*Hutnické Listy*, 1957, **12**, (9), 788–794). Breakage often occurs during fabrication or in service. Cracking tendency in welds of 347 steel stabilized with Nb is discussed and structure, welding stresses and heat treatment considered. Welds often fail by the effects of heat near the weld on the base material, σ-phase and segregations being unfavourable. Weld metal containing some ferrite is a partial solution. Quality specifications do not take these factors into account as yet.

Rapid Method for Elucidating the Primary Structure of Weld Seams. A. A. Rossoshinskii. (*Zavodskaya Laboratoriya*, 1956, **22**, (5), 558–560). [In Russian]. A simple and reliable method for studying the primary structure of weld-seam metal is described. It is based on double etching of specimens, first in 20% aqueous potassium acetate electrolytically (stainless steel plate cathode, inter-electrode distance 20–25 mm, duration 3 min) and then, after washing and drying, in a reagent consisting of 4 ml HCl, 5 g. Cu_2Cl_2, 25 ml ethanol and 30 ml water. Results obtained with a weld seam in low-carbon steel and with one containing 2% nickel after heating at 1200° C are illustrated. The method is applicable to low- and medium-alloy steels.—S. K.

On the Strength of Oblique Welds. V. Voříšek. (*Oceliarsky Sbornik*, 1954, 1955, 118–129). [In Slovak]. It is shown on the basis of tests that the weakening effect of faults in welds in tensile specimens was lowest if the weld was inclined at 45° to the tensile axis. A semi-empirical formula giving the weakening effect as a function of the inclination of the butt weld to the tensile axis, in use in the Materials Testing Institute in Zurich, was found to be suitable for the interpretation of the observed weakening. Its use in practice is advocated.—P. F.

Studies on the Oxy-Acetylene Gas Cutting of Low-alloy High-Strength Steels. K. Niguma. (*Tetsu to Hagane*, 1957, **43**, Mar., 303–304). [In Japanese]. Average cutting conditions are established, in terms of cutting speed related to plate thickness and tip angle (with and without preheat), and oxygen pressure related to tip angle. The width of the heat-affected zone is shown for the top, middle and bottom of plates up to 60 mm thick. A hardness survey is made across a typical heat-affected zone.—K. E. J.

Works Experience with the Use of Powders in Cutting with Oxygen. L. Killinger. (*Zváranie*, 1957, **6**, (7), 211–218). [In Czech]. Torches for cutting alloy steels of the Cr–Ni type with O_2 were developed. In one type, iron powder is used to facilitate cutting, in the other fine silica sand. Both methods were technologically and economically successful, though the use of silica sand necessitates stringent health precautions. The torches and their mode of operations, as well as results obtained, are discussed in detail.—P. F.

The Influence of the Impurities in Oxygen Used for Oxy-acetylene Cutting. D. Both. (*Met. Constr. Masini*, 1956, **8**, (9), 27–28). [In Rumanian]. The impurities affect the quality of the cut, lead to a lowering of the cutting speed; to an increase in consumption of gas etc. The users are advised to check the purities of the deliveries before use.

MACHINING AND MACHINABILITY

Cutting Properties and Machinability of High Speed Steels. E. P. Nadeinskaya. (*Vestnik Masinostroenniya*, 1956, **36**, (7), 35–39). [In Russian]. In view of the high popularity which a new quality of high speed steel is now enjoying in the Soviet Union, comparison tests are here described between various kinds of such steels. The experiments have been carried out with the help of active radioisotopes. The chemical composition and the mechanical properties of the steels under investigation are listed. The conclusions show that the popularity of the new steel P9F is justified.—L. H.

Vibration Stability in Boring Mandrels. L. K. Kuchma. (*Vestnik Mashinostroenniya*, 1956, **36**, (9), 33–39). [In Russian]. The basis of this study is the theory of vibration in metal cutting of Koutchymovu and Sokolowski. Theoretically the problem is first simplified by assuming only one or two degrees of freedom in the system and by accepting as a negative influence the frictional force on the cutting speed. This formed the mathematical basis of practical experiments, the results of which show conclusively that the use of mandrels with cutting discs, judiciously spaced relatively to the cutting instrument, substantially increases the vibration stability of the system.—L. H.

A Study of the Deformed Layer Produced on Metal Surfaces by Mechanical Machining, Abrasion and Polishing Operations. L. E. Samuels. (*Electroplating Metal Finishing*, 1957, **10**, Sept., 279–282, 285; Oct., 315–318, 343). Metallographic structures and the depth of the deformed layers produced in austenitic steel and non-ferrous metals by various treatments are reviewed. Production of strain-free surfaces is discussed by electrolytic or mechanical polishing methods. Machining and abrading operations are then outlined and preparation of deformation-free surfaces. Depths of scratches and deformed layers are tabulated. (38 references.)

A Study of Leaded Free-Cutting Steels. IV. Mechanical Properties of Leaded Low Ni–Cr–Mo Steel at Elevated and Low Temperatures. H. Ohashi. (*Tetsu to Hagane*, 1957, **43**, Mar., 400–402). [In Japanese].—K. E. J.

Study of Leaded Free-cutting Steels. V. Machinability of Alloy Steels and the Corresponding Free-cutting Steels. T. Araki. (*Tetsu to Hagane*, 1957, **43**, Mar., 402–403). [In Japanese]. Data are given for 22 steels (compositions given) and for gun-metal.—K. E. J.

Electro-Erosion Takes Heavier Cuts. (*Metalworking Prod.*, 1957, **101**, Sept. 27, 1755–1757). Machines seen at Hanover are described, made by Agietron.

CLEANING AND PICKLING

Coated Abrasives in Metal Finishing. J. Zoethout. (*Product Finishing*, 1957, **10**, July, 53–58, 120). An account of wheels and belts and their uses. The advantages of abrasive belts are set out.

Centrifugal Polishing—A New Polishing Method. E. Knuth-Winterfeldt and U. Langer. (*Metalloberfläche*, 1957, **11**, Aug., 253–254). Experiments are reported on the chemical polishing of steel in an oxalic acid—hydrogen peroxide bath under the influence of a centrifugal field of about 175 G. Advantages include a shorter polishing time and lower working temperature.—L. D. H.

On Shot for Peening. VI. Effect of Shot on the Cleaning Effect. K. Kamishohara. (*Tetsu to Hagane*, 1957, **43**, Mar., 239–240). [In Japanese]. Fatigue test results are given for specimens of low-alloy steel (analysis given) heat-treated; heat-treated and shot-peened; and heat-treated, shot-peened and heat-treated. In each case a comparison is made between specimens first ground and buffed, and those not so treated.—K. E. J.

Flame Cleaning. (*Torch*, **1957**, Summer, 2–8). A review of the process as pre-treatment for painting.

Theory and Practice of Flame Cleaning. K. W. Sippell. (*Werkstoffe Korrosion*, 1957, **8**, Apr., 185–216). This is a detailed survey of the principles underlying flame cleaning, and its application. The paper discusses the most suitable conditions, the selection of fuel gas, the influence of the temperatures produced on steel properties, and practical details of the application of the process. (78 references).

Effluents from Metalworking: Chrome Liquors. (*Usine Nouvelle*, 1957, **18**, Aug., 22, 73, 75). Processes yielding chrome effluents and the methods of treatment and recovery are briefly reviewed.

Neutralization of Pickling Solutions with Removal of Liquid Phase and Production of Heat-Insulating Material. V. A. Radtsig and T. L. Bruk-Levinson, (*Stal'*, 1957, (9), 858–860).

Regeneration of Waste Pickle Liquor To Produce Ferrous Sulphate Monohydrate. J. S. Atwood, J. S. Joseph and

W. W. Hodge. (*Blast Furn. Steel Plant*, 1957, **45**, Sept., 1018–1022). The process is outlined and compared economically with lime neutralization.

PROTECTIVE COATINGS

The Testing and Examination of Electrodeposits. R. Quarendon. (*Product Finishing*, 1957, **10**, Mar., 65–74, 130). **1.—Measurement of Local Thickness.** A review of jet and dropping tests, microscopical examination of cross-sections, coulometric methods and magnetic gauges is presented. **2.—Measurements of Local and Average Thickness.** (Apr., 65–73). Electromagnetic testers, thermal probes and X-ray and β-ray meters are reviewed and average thickness measurements by stripping methods are surveyed.

New Portable Device for Electrography in Works Conditions. V. P. Milin. (*Zavodskaya Laboratoriya*, 1956, **22**, (8), 997). [In Russian]. A brief account is given of a portable form of the electric instrument developed by the author for the porosity testing of electrodeposited coatings on metals and also for marking steels and other conducting alloys. The basis of the method is the production of a characteristic pattern on a special gelatin film in contact with the specimen when acting as an electrode in electrolysis.—S. K.

Metallic Surfaces Research. R. L. Samuel. (*Corrosion Tech.*, 1957, **4**, Oct., 354–356). Recent work by Metallic Surfaces Research Laboratories Ltd. on chromizing, siliconizing, refractory coatings and corrosion and heat resistance are briefly outlined.

Spraywelding. A Hard Facing Preventive Maintenance Process. J. R. Smith. (*Iron Steel Eng.*, 1957, **34**, Sept., 90–94, Discussion 94–95). A description is given of the three operations, surface preparation, spraying, and fusing the overlay. Spraying is by the powder process and the gun is mechanically operated when possible to give a uniform deposit. Fusion is carried out in a controlled atmosphere furnace, or by torch. Machining and finishing, a list of metals that can be used with and without special precautions, and a review of uses are added.

The Relationship between Temperature and Current Density of Chromium Plating Baths in the Production of Uniform Coatings. R. H. Rousselot. (*Chrome Dur.*, 1957, 41–42). An expression is deduced and its use in the plating of complex forms is shown.

Chromium Plating Development at the U.S. Naval Engineering Experiment Station. R. W. Reynolds. (*Trans. Amer. Soc. Mech. Eng. Preprint*, 1957, 57-OGP-6, pp. 9). Wear resistance for diesel cylinders is discussed and a wear test machine and plating plant are described. Commercial plating was also tested. Two porous types of plating superior in wear resistance to commercial samples were developed.

Effect of Hard Chromium on the Resistance of Steels. P. Morisset. (*Chrome Dur.*, 1957, 28–40). Chromium layers were examined for cracks, striations, H_2 content, internal stress and thickness and embrittlement by H_2 was studied. Fatigue of plated steel was investigated and practical recommendations are made.

Hard Chromium Plating at the "INFRATIREA" Mill in Oradea. G. Doczi. (*Met. Constr. Masini*, 1955, **7**, (12), 80–84). A newly introduced electrolytic Cr plating process is described in detail. The electrolyte consists of CrO_3 with sulphuric acid, and the process is carried out with an insoluble anode and the object for plating as the cathode. This plating has been found suitable for extrusion dies for non-metallic materials heated up to 500° C and for tools used in the repair shops for tractors and motors.—L. H.

Can Chrome Plate Solve Way Wear Problems? L. W. Raymond. (*Metalworking Prod.*, 1957, **101**, Sept. 27, 1743–1746). Hard chrome plate on machine tool ways reduces friction and is practically wear-proof, but its application is difficult. Firms making use of the process are mentioned.

On the Porosity of Chromium Plating. M. Bonnemay. (*Chrome Dur.*, 1957, 13–16). Equations are advanced from the results of a tracer study.

Notes on the Examination of Electrolytic Chromium Coatings on Steel by X-Ray Diffraction. A. R. Weill. (*Chrome Dur.*, 1957, 5–12). Coatings on mechanically and electrolytically polished steel were examined, it was uncertain whether slight differences observed were due to this or to variations in the steel. Residual strains may be concerned as it seemed that there might be less cracking on electropolished surfaces. Without isolation or heat treatment it appeared that an oxide was present in all coatings, apparently CrO_2 epitactic on chromium.

Selection Control and Use of Nickel Plating Baths. (*Galvano*, 1957, May, 42–46). This paper describes the installation used for Ni plating when a highly polished surface is to result. Lead fixtures are not to be used. The cleaning of articles to be plated, bath composition, wetting agents, treatment of the surface prior to Cr plating and the testing of the finished surface are discussed. (10 references.)—R. P.

Remarks on Galvanizing Salts. A. Gordet. (*Galvano*, 1957, Jan., 38–40). In answer to questions, the writer gives a critical account of galvanizing salts, with particular reference to claims made for one proprietary product. The attack on zinc is a function of temperature, immersion time, viscosity, rate of withdrawal, surface finish and also of the flux present. A thin galvanizing layer should be aimed at and only pure zinc used.—R. P.

Studies on Measurement of Zinc Coatings. O. Yanuba. (*Tetsu to Hagane*, 1957, **43**, Mar., 306–307). [In Japanese]. An electrolytic stripping method is described, and a curve shows the rise in potential as the successive phases of η, $\eta + \zeta$, ζ, δ and pure Fe are bared. Good agreement is shown with the $SbCl_3$ method.—K. E. J.

1. The Increased Attack on Iron by Zinc in the Temperature Region of 500° C. 2. Influence of the Antimony Content on the Attack on Iron by Molten Zinc. D. Horstmann. (*Forsch. Wirts. Nordrhein-Westfalen.* No. 290, pp. 30). Experiments on the transformation $\delta \rightleftharpoons \zeta$ in the Fe–Zn system showed that the formation of the ζ phase can easily be suppressed and occurs at a high rate only when strong undercooling is used. Therefore with low viscosity and little super-cooling very little ζ will form at 500° C. The δ, layer is attacked by the molten zinc at the same time, causing cracks and leading to further attack. The attack is similar in presence of Sb, the law being parabolic at low and high temperatures with respect to time. In the linear region, the temperature range 495–515° C is increased by the presence of Sb. (12 references.)—R. P.

Combined Patenting and Galvanizing of Wire. N. S. Zolotukhina, F. K. Tulenkov and I. I. Vainshenker. (*Stal'*, 1957, (2), 165–168). [In Russian]. A semi-industrial scale installation in which patenting and galvanizing of wire are carried out is described, and operating results given. Among the advantages of the procedure developed are simplicity of plant-construction and cheap operation, and its adoption in industry is recommended.—S. K.

Hot Dip Tinning of Grey Cast Iron. E. Hauszner. (*Kohászati Lapok*, 1956, **11**, (11–12), 255–261). The author reviews briefly the hot dip tinning of steel sheets and afterwards discusses fully the various hot dip tinning processes for grey cast iron, namely the half chloride, chloride, nitric and diffusion methods and their advantages and disadvantages. Finally he mentions the cold tinning processes as the plunge, contact and electrolyte methods.—P. K.

The Influence of Some Process Variables on the Reflectivity and Surface Appearance of Electrolytic Tinplate. R. Mills. (*Bull. Inst. Metal Finishing*. 1957, **34**, Autumn, 8–17). Apparatus designed to simulate the procedures used in electrolytic tinplate manufacture was used to determine the effect of baseplate finish, pickling and operating conditions. The brightness of coatings was found to be particularly affected by the surface finish of the steel base and all but the lightest pickling treatment. Increasing the temperature of the electrolyte and the relative cathode-electrolyte velocity resulted in an extension of the upper limit of current density at which acceptable coatings can be produced.—A. D. H.

New Surface Treatment Substitutes for Tinplate. (*Iron Age*, 1957, **179**, June 6, 106–108). **Treated Steel Replaces Tin Plate.** (*Steel*, 1957, **140**, June 3, 105). The "Hinac" process is described. The inorganic coating of chromium compounds on black plate is very suitable for detergent containers due to its high resistance to alkalinity. The coating also provides good corrosion resistance for other metals, and is an excellent base for enamel.—D. L. C. P.

Practical Aspects of Cadmium Plating. J. Kosmos. (*Plating*, 1956, **43**, Oct., B1235–1240). The operation, characteristics and maintenance of cyanide cadmium plating baths are discussed. Methods for chemical analysis of bath constituents are given in detail.—A. D. H.

Stainless Steel Cladding of Mild Steel and Cement Surfaces. P. Voigt. (*Acciaio Inossid.*, 1957, **24**, March-June, 48–60).

[In Italian]. The article describes a method of cladding mild steel vessels or structures and cement surfaces with stainless steel sheets. The method of fixing the sheets is patented. An important application is in the chemical industry.—M. D. J. B.

Surfacing with Stainless Prevents Pit Corrosion. J. A. Rau. (*Iron Age*, 1957, **180**, July 18, 120–121). The application of 18–8 stainless steel welding to form the surface material of vanes and shafts of extrusion presses was found to avoid trouble from surface pitting. A heavy duty lathe has been adapted to apply the weld.—D. L. C. P.

Plating of Tungsten by an Oxidation–Reduction Cycle. M. G. Charlton and G. L. Davis. (*Bull. Inst. Met. Finish.*, 1957, **34**, Autumn, 28–39). The deposition of W by an oxidation-reduction cycle in which a W source is heated (>2000° C) in an atmosphere of H_2 and water vpaour with the body to be plated placed close to the source and heated to approx. 700° C. The process is discussed and considered to afford a practical method of applying adherent coatings to ferrous materials. (28 references).—A. D. H.

A Survey of Chromate Treatments. W. E. Pocock. (*Indust. Finishing*, 1957, **9**, May, 561–568). The author gives a detailed account of the methods of application of chromate conversion coatings for the corrosion protection and brightening of galvanized, zinc and cadmium plated and non-ferrous components. The properties of the coatings are also considered.—A. D. H.

Protective Coatings with Pure Oxides. (*Metalworking Prod.*, 1957, **101**, Mar. 15, 454–455). A protective coating process (Rokide Coating Process) developed primarily for thermal purposes has proved to have promising applications where hardness and wear resistance are required. This protective coating is claimed to provide a wear-resistant surface on metals and other materials with a coating hardness comparable with the alumina used on grinding wheels.

Testing of Coal Tar Coatings—Resistance to Fouling and Degradation by Marine Organisms. R. B. Teel and W. F. Fair jun. (*Corrosion*, 1957, **13**, Aug., 491–500). Compositions resisting barnacles are described.

Bituminous-Based Protective Coatings. W. Murray. (*Corrosion Techn.*, 1957, **4**, June, 202–203). The author outlines the work of the Andrew Maxwell Division of the Liverpool Borax Co. Ltd. in the field of bituminous paints. The properties of such paints (colour, gloss retention, adhesion and corrosion resistance) are briefly discussed.—L. E. W.

Protective Coatings in Industry—Review of Modern Synthetic Materials. E. A. Suligal. (*Corrosion Techn.*, 1957, **4**, June, 198–200). The properties to be sought in a satisfactory protective coating system are listed and the extent to which they are fulfilled by modern synthetic coatings is reviewed. Chlorinated rubber paints, vinyl systems, neoprene paints, "Epikote"-resin-based paints and bitumen paints are considered.—L. E. W.

Adhesion of Phenol-Formaldehyde to Various Refractory Oxides. J. K. Sprinkle and H. F. Taylor. (*Amer. Found. Soc. Preprint*, 1957, No. 57–56, pp. 4). The experimental technique used to determine the surface tension of phenol-formaldehyde resol and the contact angle between this liquid and various refractory oxides, is described. From the results obtained the work of adhesion between the resol and each oxide is calculated from the relation $W_{AD} + \delta_L (1 + \cos \theta)$ where W_{AD} is the work of adhesion, δ_L the surface tension, and θ the contact angle. Data for the tensile strength of the resin bonded oxides are given, and it is shown that the strength is closely related to the work of adhesion between the resin and the respective oxides.—B. C. W.

"Plasteel." W. Bullough and T. A. Canning. (*Sheet Metal Ind.*, 1957, **34**, June, 431–433). A description of the B.I.S.R.A. plant and process for the production of P.V.C.-coated steel strip is given. Properties and advantages of the laminate are discussed, and it is claimed that "Plasteel" can be deep drawn, roller formed, sharply bent, punched or sheared without damage to the plastic or breaking the bond with the steel. Other P.V.C. coating processes, such as coating wire and tube, are being investigated.—A. H. M.

Mill Additions and their Effect on the Properties of Vitreous Enamel. E. Greenwell. (*Inst. Vit. Enamel. Bull.*, 1957, **7**, March, 43–48, 57). The effect of mill additions on colour uniformity, gloss, opacity, adherence, abrasions, acid and heat resistance are discussed and summarized.—A. D. H.

Relation between Roughness of Interface and Adherence of Porcelain Enamel to Steel. J. C. Richmond, D. G. Moore, H. B. Kirkpatrick, and W. N. Harrison. (*Inst. Vit. Enamel. Bull.*, 1957, **7**, May, 69–79). Porcelain enamels having varying degrees of adherence were prepared by alterations in composition of the grit or by varying the surface treatment before application of the enamel. Standard adherence tests were carried out and showed that adherence depended directly upon the roughness of the interface which was developed in part during the firing process. Pickled surfaces showed better adherence than sand blasted surfaces for the same degree of roughness. Factors other than roughness of interface also influence adherence. (18 references).

The Galvanic Corrosion Theory for Adherence on Porcelain-Enamel Ground Coats to Steel. D. G. Moore, J. W. Pitts, J. C. Richmond and W. N. Harrison. (*Inst. Vit. Enamel. Bull.*, 1957, **7**, Mar., 49–57). The theory is outlined that the adhesion of enamel occurs by mechanical anchoring into pits formed by the galvanic attack of the enamel on the steel surface. It was shown that galvanic corrosion could occur during the short firing times of enamel processing but that this mechanism was not the only factor affecting bond strength.—A. D. H.

Fifteen-Year Exposure Test of Porcelain Enamels. D. G. Moore and W. N. Harrison. (*Nat. Bur. Standards, Building Materials and Structures Report*, 148, 1957, June 28, pp. 13). Steel panes with 14 types of enamel were exposed at four stations in the U.S. Gloss, colour change and ease of cleaning are reported. All prevented corrosion where initial covering was complete.

Formation of Fishscale Defects in Enamelled Coatings Applied to Sheet Steel. Their Relationship to Certain Properties of the Base Metal. P. Tyvaert and R. Piva. (*Fonderie*, 1957, April, 153–160). The formation of fishscale defects by H_2 on the surface of enamelled steel sheet was studied as a function of the steel composition, the surface preparation given before enamelling and the hardness of the steel. In the accelerated test used H_2 was produced by acid attack at one face of the specimen and the time required for the defect to develop at the enamelled surface was measured. Of the specimens given the conventional cold sulphuric acid treatment before enamelling some showed defects in times between 35 and 105 min, and others no defects after 2 hr. The latter group had a total C, Si, Mn, S and P content between 0·071 and 0·389%, a total S and P content between 0·018 and 0·031%, a P content between 0·004 and 0·015%, and a Vickers hardness between 77 and 110.—B. C. W.

Zinc-Rich Paints Protect Iron and Steel Parts. R. J. Fabian. (*Mat. Methods*, 1957, **46**, July, 102–104). The uses of paints containing 80–95% Zn, in either organic or inorganic binders are described.

Steelworks' Maintenance Painting—15 Years' Experience at Corby. W. J. Chalmers. (*Corrosion Techn.*, 1957, **4**, June, 195–197). The author describes briefly the methods of testing new materials which show promise in small scale application. The pretreatment of surfaces, for example in the coke-oven area, is also dealt with. The author concludes by outlining the most useful paint materials; 'Drygalv' zinc-rich paint has proved a very successful anti-corrosive primer and finishing enamels prepared from synthetic resins are finding increasing application.—L. E. W.

The Use of Electrolytic Tinplate. W. R. Lewis. (*Werkstoffe Korrosion*, 1957, **8**, Aug./Sept., 456–462). The development of electrolytic tinplate is outlined, applications described, and the advantages and disadvantages compared with other processes are discussed.—L. D. H.

An Examination of Oxide Films on Tin and Tinplate. S. C. Britton and K. Bright. (*Metallurgia*, 1957, **56**, Oct., 163–168). Films formed in air at various temperatures and in passivating solutions were studied by cathodic reduction and by detaching from the surface, although films formed in chromates depend upon the Cr for chemical resistance their growth can be estimated from the tin contained. Effects of P, In and Zn are reported.

Aluminium Cladding of Steel. (*Ajour Tek. Inf.*, 1957, **10**, July, 117). [In Norwegian]. The article briefly describes a method of cladding steel with Al, Ti or Mo for preventing corrosion and also to render it fit for certain heavy-duty purposes, notably heat resistance and gas exhausts. The method itself has been developed by the Swedish engineer G. A. Moeller and is known as moellerizing.—F. R. H.

Titanium-clad Steel Sooner Than You Think. R. F. Domagala, D. W. Levinson and W. Rostoker. (*Product Eng.*, 1957, **28**, Sept. 16, 111–113). A progress report on roll-cladding. The problem of intermetallic compound formation, the need for an intermediate layer and microstructures of interfaces are discussed.

An Iridium Plating Solution. P. J. Ovenden. (*Product Finishing*, 1957, **10**, Mar., 62). The solution was developed for plating on steel wire with or without a Cu undercoat.

Standard Specifications for Coal-Tar Enamel Protective Coatings for Steel Water Pipe. American Water Works Association. (*J. Amer. Water Works Assoc.*, 1957, **49**, July, 941–968). AWWA C203–57, approved Aug. 15, 1957 is given.

Corrosion-Resistant Coatings. E. G. Rawlings. (*Corrosion Tech.*, 1957, **4**, Aug., 283–286). Rubber and its derivatives are reviewed. Natural and synthetic rubbers and chlorinated rubber are briefly considered.

Silicone Protective Coatings for the Steel Industry. R. C. Hedlund. (*Iron Steel Eng.*, 1957, **34**, Sept., 109–112). The use of silicone and silicone-synthetic resin paint and varnish coatings is reviewed. Their water repellency was referred to in discussion.

Enamel on Steel. G. Minoux. (*Acier-Stahl-Steel*, 1957, **22**, Mar., 111–112). Steel enamelled for decorative effect is used on shop fronts and in other locations.

New Leaded Porcelain Enamels for Steel Make Possible Fresh Fields of Application. (*Lead*, 1957, **21**, (2), 3). A brief account is given of new enamels with low firing temperatures (1000–1100° F).

Enamel Firing Ovens of the Lysva Works. G. Ya. Ustinov and G. S. Melamud. (*Stal'*, 1957, (9), 847–849).

POWDER METALLURGY

Powder Metallurgy as an Aid to Production Engineering. G. R. Bell and L. M. Marucchi. (*Engineer Foundryman*, 1957, **22**, Aug., 54–58; Discussion, 59–60). A review of methods, materials, inspection, control of porosity and finish, factors in the design of parts, impregnation, stainless steels and non-ferrous metals, with notes on uses.

Höganäs Iron Powder Handbook. Vol. I—Basic Information on Iron Powder Metallurgy. (1957, pp. 150). A brief history is followed by a survey of methods of production, of the physical properties of powders and of sintered test bars. Technical uses are then reviewed and thermal treatment, lubrication, alloying additives, pressing and tooling, compacting, sintering and sintering atmospheres and design of parts, joining, peen plating and the layout of pilot and full-scale plant are all dealt with. The volume is loose-leaf, and further chapters are to follow.

FERRITES, CERMETS, AND CARBIDES

A Theory of Uniaxial Anisotropy Induced by Magnetic Annealing in Ferrites. S. Taniguchi. (*Sci. Rep. Res. Inst. Tohoku Univ.*, 1957, **9**, June, 196–214).

Paramagnetic Study of the Ferrites 5 Fe_2O_3, $3M_2O_3$ Where M = Gd, Dy, Er. R. Aléonard and J. C. Barbier. (*Compt. Rend.*, 1957, **245**, Aug. 19, 831–834).

Cermets: II, Wettability and Microstructure Studies in Liquid-Phase Sintering. N. M. Parikh and M. Humenik, jun. (*J. Amer. Ceram. Soc.*, 1957, **40**, Sept., 315–320). Surface energy relationships appear to control the process and a general "coalescence hypothesis" is advanced.

On the Systems UC–TiC,–ZrC,–VC,–NbC,–TaC,–Cr_3C_2,–Mo_2C and –WC. H. Nowotny, R. Kieffer, F. Benesovsky and E. Laube. (*Monatshefte für Chemie*, 1957, **88**, reprint, 336–343).

PROPERTIES AND TESTS

Quantitative Evaluation of the Non-Uniformity of Product Properties from the Results of Small Selections. L. P. Vladimirov. (*Zavodskaya Laboratoriya*, 1956, **22**, (8), 955–959). [In Russian]. A brief account is given of the statistical principles involved in the estimation of, e.g., the mechanical properties of a population from the results of tests on a small number of samples.—S. K.

Surface Finish: What it is: How it is Measured. R. E. Reason. (*Product Eng.*, 1957, **28**, Sept. 16, 77–81). Electronic and piezoelectric techniques for measuring and characterizing surface irregularities are reviewed.

The Importance of the Testing of Materials. Present-Day Development in the Construction of Machines and Apparatus for Testing, and for Mechanical and Technological Examinations of Metals in Eastern Germany. W. Wiesner, (*Met. Constr. Masini*, 1956, **8**, (7), 3–12). This article is a detailed review of metal testing in Eastern Germany under every heading, and is meant to be a guide for the Rumanian Industry.

The Centre of a Dislocation: I. E. H. Yoffe. (*Phil. Mag.*, 1957, **2**, 8th Series, Oct., 1197–1210). An attempt is made to describe mathematically the state of stress at the centre of an edge dislocation.

On the Origin of Elastic Defects in a Polycrystalline Alloy. A. B. Gur'ev. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (2), 349–359). A new proof is given that elastic hysteresis loops are formed as a result of a local shear in several micro volumes of the metal under elastic deformation of the bulk of the grains. It is proposed to use a new constant, the coefficient of micro-plasticity, to describe the tendency of nuclei to elastic flaws. Results of experimental study of this constant are given.—L. H.

The Influence of Elastic Stresses and of Thermo-Mechanical Treatment, on the Magnetic Properties of Some Permanent Magnetic Materials. M. G. Luzhinskaya and Ya. S. Shur. (*Fizika Metallov i Metallovedenie*, 1957, **4**, (2), 239–244). [In Russian]. A comparison of the heat treatment and of elastic stresses on the magnetic properties of the alloys Fe–Co–V, with the alloy containing 12%V 52% Co and remainder Fe, investigated previously, show that these influences are similar. The results obtained justify the conclusion that the physical causes of the change of properties under the influence of stress, become superimposed on the changes during heat treatment. The conclusion is evident that this heat treatment influences the magnetic structure and therefore also the magnetic properties, and shows itself as one of the modern methods of increasing such magnetic properties.

Changes in the Plasticity of Transformer Steels During the Cooling Process. S. I. Doroshek, N. I. Lapkin and G. N. Shubin. (*Fizika Metallov i Metallovedenie*, **4**, (1), 171–176). [In Russian]. The viscosity of these high-alloy transformer steels during the cooling process after annealing, follows a curve showing a maximum. This law becomes evident in stepped cooling. A speeding up of the cooling process after annealing increases the viscosity. The presence of the coarse cementite, even along the boundaries of the ferrite grains, although it increases the brittleness of the steels, is not the only cause of it.—L. H.

A Method for Waterproofing Bonded Electrical-Resistance Strain Gauges. R. E. Morris and R. R. James. (*J. Amer. Soc. Naval Eng.*, 1957, **69**, Aug., 527–530). A neoprene dam filled with mastic is used.

Tensometric-Sensitivity of a Conductor and an Adhering Wire in a Plane Strained State. E. Yu. Nekhendzi and N. G. Tisenko. (*Zavodskaya Laboratoriya*, 1956, **22**, (8), 982–989). [In Russian]. The range in which the tensometric-sensitivity coefficient remains constant has been determined for a variety of wire resistance strain gauge systems. From this and a consideration of the strained state of the specimen-gauge system a method of placing strain gauges is recommended.

Comparison of the Properties of Cast Iron under Static Tensile and Flexural Strain. P. le Rolland and E. Plenard. (*24th International Foundry Congress Papers*, 1957, 159–168). Elasticity under various conditions of stress is considered. Extrapolation to zero stress is studied and the slope is shown to depend on damping capacity and structure of the (grey) iron tested.

Fractography Applied to the Detection of Defects in Cast Steel. K. I. Basov. (*Liteinoe Proizvodstvo*, 1957, (1), 16–17). [In Russian]. The advantages of fracture methods for establishing the causes of failure of steel castings are mentioned and a proposed variant of the method is briefly described. The method consists essentially in cutting a test piece from the section examined, subjecting it to hardening and tempering, (600–650° C), and then breaking it under static load after notching. The comparatively low plastic properties of unsound portions distinguishes them in the fracture. Sometimes the same effect can be obtained with specimens in the normalized or annealed states.—S. K.

Hooke's Law and the Concept of Elastic Limit. E. Williams. (*Annals of Science*, 1956, **12**, March, 74–83, Reprint). Historical.

Damping Characteristics and Decay of Vibration Amplitude of Metals and Alloys from Energy Considerations. S. Ghosh, K. C. Som, A. K. Lahiri, and G. P. Chatterjee. (*Trans. Indian Inst. Met.*, 1954–55, **8**, 165–180; discussion, 180–185). The construction of a simple instrument for the measurement of damping capacity is described and values are presented for carbon and low alloy steels. A general relation is developed between the amplitude decay characteristic and the energy lost in the stress cycle assuming that metals behave neither perfectly elastically nor plastically.—A. D. H.

Dislocations and Etch Pits in Iron. (*Canad. Mines Tech. Surveys, Phys. Met. Division*, (July–Dec. 1956), 1957, Jan. 15, 13–15).

Design Properties of High-Strength Steels in the Presence of Stress Concentrations. B. B. Muvdi, E. P. Klier, and G. Sachs. (*PB* 121155, 1956, *T.I.D.U. List* 750, pp. 105). Tension, notch-tension, impact, fatigue, and notch-fatigue tests are tabulated as a supplementary report.

Investigation of Stress Concentration in " Admiralty " Test Pieces. B. A. Meisner. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 853–855). [In Russian]. An optical polarization study has been made of stress concentration in models of eccentrically tensile loaded plate test-pieces, used for the brittle fracture testing of steel. Results for models on different scales are shown.—S. K.

The Effect of Phosphorus on the Tensile and Notch-impact Properties of High-purity Iron and Iron–Carbon Alloys. B. E. Hopkins and H. R. Tipler. (*J.I.S.I.*, 1958, Mar., **188**, 218–237). [This issue].

Type MSh-70 " Ukrinmet " Machine for Testing Thin Metal Sheets by the Biaxial Tension Method. A. K. Khodulin and N. I. Sandler. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1241–1244). [In Russian]. A testing machine for sheets up to 2–3 mm thick of high quality carbon cold-rolled steel is described. The machine has been used under industrial conditions for over 1500 specimens of a variety of steels. It uses biaxial tension and is said to have a wide applicability.

Comparison of Mechanical Characteristics Obtained with Flat and Round Test-Pieces. P. R. Rusinov. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1225–1227). [In Russian]. Tests on a ship-plate steel using test-pieces of different shapes showed that this factor has practically no effect on values for yield point or ultimate strength.—S. K.

The Use of Notched Test Bars in Materials Testing. K. Lotsch. (*Maschinenbau Wärmewirtschaft*, 1956, **11**, Sept., 252–257). A review of theory is presented.

The Yield-stress Temperature Relation for Iron at Low Temperature. J. S. Erickson and J. R. Low jun. (*Acta Met.*, 1957, **5**, July, 405–406). The low temperature deviation from the expected curve is attributed to the intervention of twinning as the initial mode of deformation and the suppression of the yield-point phenomenon.

Method of Determining Resistance to Breaking of Plastic Metals. Yu. I. Likhachev. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1209–1217). [In Russian]. A method has been developed for finding the approximate resistance to breaking of steel and other plastic metals in the cold-worked and unworked states, under static, uneven, tri-axial tension at ordinary temperatures. Ring-notched test pieces are used. For metals with resistance to breaking considerably in excess of the yield strength the method enables the lower limit of the resistance for the cold-worked metal to be found. The strained and deformed state at the minimal cross section of a steel test-piece has been analysed. The effect of cold-working on the resistance to breaking has been estimated.

The Third Period of Stress Relaxation in Metals. V. Z. Tseitlin. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 845–849). [In Russian]. The results are presented of prolonged investigations on the relaxation of a large group of steels and alloys of various types used in steam turbines, and previously obtained relaxation curves are analysed. Materials dealt with included pearlitic-type carbon and low-alloy steels, carbide-type steels, iron–chromium–nickel and iron–chromium–nickel–manganese austenitic steels (including those prone to σ-phase formation), iron–chromium–nickel–cobalt, and nickel base alloys. Most tests continued for 600–10,000, some for 20,000 hours, at various temperatures and initial stresses. In no case was any " third stage " of relaxation (supposed to be associated with growing creep rate) observed.—S. K.

Simplification of Relaxation Tests on Metals and Alloys. Ya. S. Gintsburg. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 840–845). [In Russian]. The possibility of constructing relaxation curves for any stress from curves obtained with two or three values of initial stress under constant temperature conditions is examined. Special attention is given to relationships for the second stage of relaxation, examples of curves obtained for steels and other alloys with simple and logarithmic coordinates being shown. Relaxation equations are deduced, values of the empirical coefficients for carbon, low-alloy and austenitic steels heat treated in known ways being tabulated. The use of these data for the simpler evaluation of relaxation characteristics is illustrated.—S. K.

Brimor High Temperature Cements. Morgan Crucible Co. (*Technical Information*, (1957), pp. 4). The materials were developed to secure strain-gauges to steel and operate up to 750° C.

The Influence of Local Plastic Deformation on the Tensile Strength of Steel. A. E. Asnis. (*Avtomaticheskaya Svarka*, 1956, (3), 88–94). [In Russian]. Local plastic deformation associated with cutting and punching reduces the tensile strength of Bessemer steel and O.H. killed steel much more than that of O.H. rimming steel. Work hardening caused by punching causes a greater reduction in damping capacity of Bessemer steel than in that of O.H. killed steel. Both cutting and punching increase initial hardness by 40–70%. Tensile properties can, however, be almost completely restored by normalizing. The experiments showed that during welding, the temperature of 900° C and above is distributed at a distance of 4–5 mm from the edge of the bead.—R. S.

A Differential Torsion Testing Machine. N. H. Cook. (*Proc. Soc. Exper. Stress Anal.*, 1957, **14**, (2), 155–158). Two specimens of slightly different size are tested together in the machine described, thus giving stress–strain data which are not ambiguous as to strain rate.

Method for Recording True Curves of the Resistance of a Metal to Compression. G. E. Arkulis. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1217–1220). [In Russian]. In the method described the cross-sectional area of a cylindrical compression test-piece is kept constant during the test. This is achieved by the continuous removal of the " barrel " by a cutter rotating about the test-piece axis with a radius equal to that of the undeformed test-piece.—S. K.

Notch Ductility of Commercial Malleable Irons. H. F. Bishop, G. A. Sandoz, N. C. Howells, and W. S. Pellini. (*PB* 111999, 1956, *TIDU List* 750, pp. 25). The drop-weight and explosion crack-starter tests are applied to ferritic and pearlitic irons. In the former a max. temp. is established below which brittle fracture will occur in presence of a crack, this is called Nil Ductility Temp. Results are compared with Charpy-V curves.

Studies on Shock-resistance of Steel. I. On Toughness and Microstructure in Heat-treated Tool Steel. M. Tanaka. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Sept., 509–513). [In Japanese]. Comparisons are made between a die steel and a specially-prepared Si–Cr–V steel; the toughness values were found by the static bending test. Differences between the specimens, or in specimens, even when hardnesses were approximately the same, are attributed to variations in the amounts of solute in the martensite or austenite, and the shapes of decomposition products in the matrix. The second steel is recommended as a shock-resistant tool steel. (12 references).—K. E. J.

Potentialities of the Impact Machine for Producing High Accelerations. M. Kornhauser. (*Proc. Soc. Exper. Stress Anal.*, 1957, **14**, (2), 21–32).

How to Impact Test Sheet Material. R. F. Domagala and H. L. Nurnberg. (*Iron Age*, 1956, **178**, Nov. 15, 136–137). A simple arrangement set up by the Armour Research Foundation using a standard Sonntag tester is described. Hot specimens can be used.

The Effect of Work Hardening the Edge of Sheets in relation to the Tensile Strength of Lap-Welded Joints. B. F. Lebedev. (*Avtomaticheskaya Svarka.*, 1956, (3), 95–101). [In Russian]. The cutting of sheets by means of shears gives rise to marked work hardening and the formation of tears at the edge of the metal, thus reducing the static tensile strength and causing brittle failure of the longitudinal lap-welded joints. The degree of reduction of strength and ductility of such lap-welded joints increases as the testing temperature is lowered and with an increase in sensitivity to stress concentration and susceptibility to ageing of the base metal.—R. S.

Studies on Helical Springs Made of Twisted Wire. III. Steel Wire Springs. IV. Helical Springs of Stainless Steel, Cu Alloy and Light Alloy Wires. S. Mito. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Sept., 529–532; 533–536). [In Japanese]. III. The effect of pretwisting on helical tension springs of cold-drawn steel was to increase the elastic limit by 50–100% and the fatigue limit by 20–40%, and to reduce the spring constant (modulus of rigidity) several per cent. The optimum blueing temperature was hardly affected. (10 references). IV. Similar results are reported for stainless-steel, phosphor-bronze, brass, and Duralumin wires.—K. E. J.

On the Effect of Working Velocity on the Stored Energy of Cold-worked Armco Iron. G. Mima and T. Yamane. (*Tetsu to Hagane*, 1957, **43**, Jan., 28–31). [In Japanese]. The specific heats of specimens cold-compressed by a high-speed working velocity (4·3 m/s) and a low-speed one were measured, and the stored energies compared. The stored energy was released in two steps. The ratios of stored to working energy were 4·9% in the first case and 9·4% in the second. The stored energies increased in proportion to the ratio of height of the uncompressed to the compressed specimen.—K. E. J.

Fatigue Tests on Notched Mild Steel Plates with Measurements of Fatigue Cracks. N. E. Frost and D. S. Dugdale. (*J. Mech. Phys. Solids*, 1957, **5**, (3), 182–192). Conditions for the production of no cracks, of non-propagating cracks and of complete failure are determined in terms of root radius, mean stress, and stress range.

The Fatigue and Damping Capacity of Metals and Their Dependence on Surface Hardening by Cold Rolling. A. Vetiška. (*Hutnické Listy*, 1956, **11**, (11), 648–654). [In Czech]. Experiments on the damping capacity and fatigue strength of $\frac{1}{2}$ in. dia. rods of a 1·4% Si and 0·55% Cr steel, hardened to give a tensile strength of 19 t/in², are reported. The fatigue tests were carried out on a torsion machine. The fatigue strength was found to increase with the degree of cold work; the damping capacity showed a similar trend. The effects are analysed theoretically.—P. F.

Determining the Fatigue Limits in Alternating Torsion of High-Strength Wires of Small Diameters. K. Éliáš, F. Kotrbatý, and F. Pokorný. (*Hutnické Listy*, 1956, **11**, (12), 728–732). [In Czech]. The fatigue limits of 0·74 to 0·91% C carbon steel wires of 1·5 to 2·1 mm dia. were determined by means of a Schenk-Webi torsion machine. The dependence of the fatigue limit on the wire diameter is discussed, and data from similar tests obtained by other workers are critically reviewed. The effect of surface quality and composition on the S–N curves are also considered.—P. F.

Fretting Fatigue. "Recorder." (*Metal Ind.*, 1957, **90**, May 24, 444, 446). A review with reference to non-ferrous metals.

Fatigue Properties of Welds. J. G. Whitman. (*Sheet Metal Ind.*, 1957, **34**, July, 529–538). A chapter from a forthcoming book.

Prot Fatigue Study of an Aircraft Steel in the Ultra High Strength Range. P. W. Ramsey and D. P. Kedzie. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 2, 401–406). The authors have studied the relationship between fatigue and tensile properties in a Ni–Cr–Mo–V ultra-high-strength aircraft steel, using the Prot method of fatigue testing. A useful increase in fatigue strength is found up to 240,000 lb/in² tensile strength in carefully ground specimens, with an endurance ratio of 0·40–0·42. The method offers a useful testing technique, correlating well with the Wohler method but with worthwhile savings in specimens and test time.—G. F.

Hydrogen Embrittlement and Static Fatigue in High Strength Steel. R. D. Johnson, H. H. Johnson, W. J. Barnett, and A. R. Troiano. (*PB* 121064, 1955, Aug., *TIDU List* 750, pp. 34). Static fatigue became significant at levels above 180,000 psi and the temp. dependence of delayed fracture time showed crack propagation to be controlled primarily by H_2 diffusion inward, another influence was indicated from the change of slope of the curve as a function of test temp. The embrittlement produced by Cd plating was compared with standard H_2 charging. On ageing Cd plated specimens the H_2 distribution changed from surface to uniform concentration at a level sufficient to produce embrittlement.

Creep of Mild Steel. T. Nishihara, S. Taira, and K. Tanaka. (*Tech. Rep. Eng. Res. Inst. Kyoto Univ.*, 1956, **6**, Aug., 99–125). [In English]. The creep behaviour of mild steel was studied under conditions of constant tensile stress, instead of constant load. Rheological analysis of the creep strain produced an empirical formula representing the stress–strain–time relationship. Using this, tension test results can be correlated with tension creep test results; it also allows qualitative predictions to be made of bending creep and creep behaviour under combined stresses.—K. E. J.

Creep of Low Carbon Steel under Interrupted Stressing. T. Nishihara, S. Taira, K. Tanaka, and K. Ohji. (*Tech. Rep. Eng. Res. Inst. Kyoto Univ.*, 1956, **6**, Aug., 127–145). [In English]. Results for low-carbon steels at 450° C under stepwise changing stress show various cases of increased strain or strain rate, as compared with constant-stress results. These investigations more nearly represent conditions found in machinery or other parts in which creep is important. (24 references).—K. E. J.

High Temperature Creep of Coil Springs. C. T. Eakin. (*Product Eng.*, 1956, **27**, Nov., 186–190). Piano wire and 304 stainless were tested at 20°, 150°, and 250° C in apparatus shown and the creep-stress curves are shown. Creep was found to be almost an exponential function and the curves could be extrapolated to higher values on log. plots.

Embrittlement in Creep-Rupture Tests. (*Canad. Mines Tech. Surveys, Phys. Met. Division*, (*July–Dec.* 1956), 1957, Jan. 15, 16). Embrittlement is attributed to impurities.

Use of Creep to Determine the Sum of the Principal Stresses in Two-Dimensional Problems. A. J. Durelli and W. F. Riley. (*Proc. Soc. Exper. Stress Anal.*, 1957, **14**, (2), 109–116). A photoelastic method of separating the stresses is described. Changes in thickness during recovery are measured.

The Shape of Sink-in Indentations of Hardness Measurements. B. S. Satyanarayana. (*J. Sci. Ind. Res.*, 1957, **16B**, May, 222–223). Interference studies of Vickers indentations are reported.

On the Hardness of Martensite. Supplement on the Relation between Phase Diagram and Hardness. II. J. Matsuoka. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Sept., 562–566). [In Japanese]. Measurements of change of hardness and microstructure of Fe–Ni alloys due to heat treatment showed the contributions made to the hardness of martensite by thermal and quenching stresses, structure and supersaturation of carbon. Some principles necessary in inferring hardness from phase diagrams are deduced. (12 references).—K. E. J.

Hardness of Graphite Inclusions. A. N. Rozanov. (*Liteĭnoe Proizvodstvo*, 1956, (3), 12–13). [In Russian]. In the investigation briefly described the micro-hardness of graphite inclusions in the following materials was determined: Mg-inoculated cast iron; hypereutectic grey irons melted in a vacuum furnace or a graphite-element furnace from iron powder and graphite; technical grey iron; and annealed malleable iron.—S. K.

Chipping of Hand Hammers in Service. A. H. Burn. (*Engineering*, 1957, **184**, July 12, 48–49). Accidents from hammers with a hardness above the maximum of B.S. 876: 1949 are reported and dangerous chip formation from a number of specimens is described.

Theory of Precipitation Hardening: Isotropically-Strained System. R. O. Williams. (*Acta Met.*, 1957, **5**, July, 385–392). Calculations show that strengthening can result from the interaction between edge dislocations and isotropically strained matrix and precipitate resulting from a certain type of precipitation. The theory is applied to the Fe–Cr system.

On the Secondary Hardening of Low-Alloy Boiler Steels. V. Čihal and J. Ježek. (*Hutnické Listy*, 1956, **11**, (11), 669–670). [In Czech]. By means of transmission electron diffraction very finely dispersed V_4C_3 was found to be present in a 0·08% C, 0·26% V boiler steel during tempering already prior to reaching maximum hardness. The results are shown to provide support for the theory of dispersion hardening and the existence of an optimum particle size which yields maximum hardness.—P. F.

Some Estimates of the Thermal Stability of Dispersion-Hardened Alloys. A. W. Cochardt. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 2, 434–437). The author discusses the mechanism of over-ageing in precipitation-hardened alloys, and estimates the time in which structural changes should occur in alloys hardened with Ni_3Al, Al_2O_3, Fe_3C, TiC, and TiN.—G. F.

Evaluation of the Sensitivity of Steel to Temper and Thermal Brittleness. P. B. Mikhailov-Mikheev. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 849–853). [In Russian]. Some methods used for evaluating the tendency of a steel to develop brittleness through tempering or through prolonged soaking at

somewhat lower temperatures are critically discussed. Data are presented to show that the evaluation should not be based on toughness values for a single temperature but on a comparison of temperature curves of toughness for the given steel in the brittle and ductile states. These families of curves should embrace all relevant thermally-governed factors (cooling rates, tempering temperature, and time). With standard test pieces corrections for shape and size become unnecessary.—S. K.

Low Temperature Brittleness. A. W. Magnusson and W. M. Baldwin jun. (*J. Mech. Phys. Solids*, 1957, **5**, (3), 172–181). An investigation by tensile tests on iron and other metals is described and the findings discussed theoretically.

Brittle Fracture of Mild Steel. L. Aitchison. (*Iron Coal Trades Rev.*, 1957, **174**, May 31, 1273–1275). A review is given of the mechanism, so far as it is known, and suitable alloying for minimising the condition is indicated.

Flaking of Heavy Alloy Steel Sections. C. R. Garr and A. R. Troiano. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 2, 445–449). Using an experimental heat of Cr–Ni–Mo steel with abnormally high H_2 content, the authors have produced flake-free forgings by means of an anti-flaking cycle which is primarily an isothermal transformation in the pearlite range. Such forgings, however, could be flaked severely by subsequent heat treatment. Evidence is presented showing the profound effect of transformation stress on propensity to flaking.—G. F.

Effect of Speed of Cooling on Formation of Solidification Cracks. I. K. Pokhodnya. (*Avtomaticheskaya Svarka*, 1955, (6), 64–73). [In Russian]. The relationships between the momentary rates of cooling of the solidifying weld metal (C 2·1, Cr 12·5, Si 0·3, Mn 0·77, W 1·1, V 0·2) and the temperature of heating prior to welding and the power consumption per foot run were examined. The distribution of cooling speeds in the weld pool were investigated. It was found that the optimum speeds of cooling to avoid hot cracking varied with differing conditions of welding. An increase in the cooling rate of the weld metal increased the amount of hot cracking. The rate of cooling can be reduced by prior heating of the part and by increasing the power per foot run.—R. S.

Mechanism of Crack Formation. V. F. Loshkarev. (*Stal'*, 1956, (8), 734–737). [In Russian]. The inability of the hydrogen hypothesis to explain all the features of crack formation in steel is indicated. Some of the author's calculations are criticised in an Editorial appendix.—S. K.

Mechanism of Crack Formation in Steel. V. S. Mes'kin. (*Stal'*, 1956, (8), 727–734). [In Russian]. From a critical examination of experimental and theoretical material the author arrives at the following main conclusions on the formation of cracks in steel: the cracks are the result of local dynamic brittle fracture of the metal produced by the effect of rapid localized heat evolution by the combination of hydrogen atoms to form molecules; an essential condition for crack formation is the anisotropy of properties which always results from plastic deformation; the hypothesis advanced enables the influence of various factors on crack formation to be assessed and measures for its minimization to be determined.—S. K.

Studies on Cementite Cracks of High-carbon Steel. III. T. Uesugi and Y. Watanabe. (*Tetsu to Hagane*, 1957, **43**, Feb., 137–142). [In Japanese]. Cracks produced by special kinds of carbide, e.g., large carbide produced by the eutectic reaction, net carbide produced at the Acm transformation and plate carbide precipitated at the A_1 transformation, are studied; they considerably reduce rupture strength. Net carbides produce different cracks with compressive or tensile stresses. Compressive stress produces unusual (e.g., " burst of carbide ") cracks.—K. E. J.

Crack Propagation in the Hydrogen-Induced Brittle Fracture of Steel. W. J. Barnett and A. R. Troiano. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 2, 486–494). The authors have developed an electrical resistance method to permit analysis of crack growth kinetics, and have identified the mechanism of static fatigue fracture of high strength steels as a process of hydrogen-induced crack initiation and propagation, followed by catastrophic fracture. Three distinct stages of crack growth have been observed in specimens with high H_2 concentrations in the surface layers; the first stage is characteristic of crack growth through a pre-existent hydrogen-rich region, whilst the second is associated with the gross inward diffusion of hydrogen and the third with imminent fracture.—G. F.

Method of Evaluating the Occurrence of Hairline Cracks in a Heat. G. S. Chernyak. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1182–1185). [In Russian]. The results are presented of a statistical analysis of hairline-crack test results on numerous samples taken from heats of steel produced in a 20-ton electric-arc furnace. The use of magnetic flaw detection is recommended.—S. K.

On the Problem of Flaky Fracture. I. S. Kaptyug. (*Stal'*, 1956, (11), 1021–1025). [In Russian]. The author considers that much of the published data on the occurrence of flakiness in fractures needs careful reassessment. He gives examples of such fractures due to various causes and suggests improved testing procedures.—S. K.

Detecting Flaky Fracture in Steels. N. S. Fedotenko. (*Zavodskaya Laboratoriya*, 1956, **22**, (7), 855–856). [In Russian]. Certain heats of two types (40 Kh and 35 Kh) of chromium tool steel were found occasionally to develop a typical flaky fracture. The existing testing technique for this defect (heating test pieces to 840–860° C in a salt bath and then cooling them in oil or nitrate at 150–200° C before fracturing) gave such indefinite results that faulty steels were liable to escape detection. Tempering for 1 h at 450–500° C after the quenching was found greatly to improve clarity. Typical fractures obtained with the two methods are illustrated.—S. K.

Abrasion Resistance. (*Electroplating*, 1957, **10**, April, 109–113). Methods of measuring abrasion resistance are reviewed. Blasting tests, in which a controlled flow of abrasive particles is maintained over the surface of the test piece and rubbing tests in which the test surface is rubbed with an abrasive wheel (the Taber Abraser) are considered in detail.—A. D. H.

Resistance to Wear of Sulphur-Rich Cast Iron. A. D. Ushakov. (*Liteĭnoe Proizvodstvo*, 1956, (4), 19–20). [In Russian]. A metallographic, physical and mechanical investigation of sulphide-treated cast irons (0·47–0·85% S) is briefly described which has shown that sulphide inclusions produce a favourable effect on the ability of cast iron bearings to get run in and on their anti-friction properties.—S. K.

Evaluation of the Wear Properties of Rails and Tyres on the Basis of Laboratory Tests. S. Hořejš. (*Hutnické Listy*, 1956, **11**, (12), 721–728). [In Czech]. An improvement of the wear test employed by the Institute of Materials Testing in Zürich, in which an Amsler machine is utilized, has been devised by the author. It is shown that the new method leads to results which are in substantially better agreement with the amount of wear observed in practice than the criteria used by the Swiss EMPA.—P. F.

Measuring the Abrasiveness of Carbonaceous Deposits. (*Sci. Lubr.*, 1957, **9**, July, 16–17). A device for making tests on pistons and cylinder liners is described.

Studies on Gas Carburisation. On the Erosion Resistance of Many Kinds of Steel and Electric Heating Elements for Controlled Atmospheres. M. Kawakami, M. Komuro, and S. Yuuki. (*Tetsu to Hagane*, 1957, **43**, Jan., 54–61). [In Japanese]. Comparative studies are made on heat-resisting steel, heating elements etc. used in gas retorts, muffles, conveyors, fans and other equipment of controlled-atmosphere furnaces for gas carburizing, bright heating and carbonitriding. Erosion caused by high-temperature oxidizing, carburizing, nitriding, etc. is classified into four categories.—K. E. J.

Non-destructive Automatic Test Methods for Steel and Semi-finished Steel. E. Hanke. (*Neue Hütte*, 1957, **2**, Feb.-Mar., 111–114). Methods using the effects of lines of force and the inductive effects of alternating fields are described. The former include the fracture detection method with Förster probes, the latter the multitest, magnatest, and ferrograph methods.—L. J. L.

A Method for the Precise Measurement of Permittivity of Sheet Specimens. A. C. Lynch. (*Proc. Inst. Elect. Eng.*, 1957, **104**, July, B, 359–362).

Use of a Ponderomotive Magnetometer in Metallography. V. A. Tolomasov. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1189–1194). [In Russian]. The disadvantages of existing ponderomotive magnetic devices are enumerated and an improved design of magnetometer is described. This combines the advantages of a ballistic galvanometer and an anisometer and enables the magnetization of both ferromagnetic and paramagnetic materials to be measured. Continuous recording of readings and the smallness of specimens

enable metastable phases to be investigated. The instrument can deal with very small quantities of a phase in the specimen and it can be used differentially. Curves obtained for two steels and Armco iron for various temperatures are shown and their significance discussed.—S. K.

Studies on Anisotropic MK Permanent Magnet. II. On Heat Treatment and Structure. T. Mishima and N. Makino. (*Tetsu to Hagane*, 1956, **42**, Dec., 1127–1132). [In Japanese]. The magnetic properties of the alloy are sensitive to heat treatment, and optimum conditions have been established. Specimens are solution-treated for 10–20 min at 1200° or 1250° C, and then cooled in a field of some 1500 oersteds; the optimum cooling rate is 1° C per sec between 900° and 800° C. The field is effective from the Curie point to 150° C below. In age-hardening, the specimen is held at 600° or 625° C for 1–3 hours; growth of columnar crystals improves the properties. Precipitation cannot be discerned in these specimens. The magnetic anisotropy and high coercive force may be caused by the sub-division of the material by needle-like pseudo-precipitations, directed by the cooling process and developed during ageing. **III. On the Magnetic Properties and Industrial Applications.** (1957, **43**, Jan., 43–47). The magnetic properties of the material are listed, and results of investigations into demagnetizing by various means, and leakage factors, are reported. The magnets are highly suitable for industrial applications, e.g. loudspeakers and magnetrons.

Experience Gained with Crack-Depth Meters. L. Brand. (*Stahl u. Eisen*, 1957, **77**, May 2, 576–581). The depth of a crack in a metallic conductor is related to variations of an electric current measured at the surface (see Trost, *Metallwirtsch.*, 1944, **23**, 308). Trost used D.C. but apparatus employing A.C. were recently put on the market. Both the D.C. and the A.C. apparatus do not serve for detecting cracks but for measuring cracks found by other means. The characteristics of the two types of meter are described. The various factors that affect accuracy in the determination are enumerated; it lies within ± 10–15% The results determined with the A.C. apparatus are independent of the depth of the crack, whereas those obtained with the D.C. model are scattered substantially for cracks deeper than 4 mm.—T. G.

High-frequency Determination of Ferromagnetic Metals. F. J. Schmidt. (*Anal. Chem.*, 1957, **29**, July, 1060–1062). Reduced oxides are examined by magnetic permeability methods and by empirical calibration, ferrous metals can be determined in a few minutes.

Autosonics. (*Metallurgia*, 1957, **55**, May, 235–238). Automatic ultrasonic inspection of solid metals is discussed with particular reference to automatic equipment developed by Kelvin and Hughes Ltd.—B. G. B.

Automatic Ultrasonic Inspection. (*Brit. Eng.*, 1957, **39**, June, 402–403). An account, with illustrations, of the Autosonics equipment.

Can Ultrasonics Speed Fatigue Testing? (*Metalworking Prod.*, 1957, **101**, June 14, 1032–1033). Apparatus developed by Mullard Research Laboratories is described.

Method for Automatic Continuous Testing of Materials with the Aid of a Low-Frequency Ultrasonic Flaw-Detector. D. F. Vasil'ev and N. F. Shustov. (*Zavodskaya Laboratoriya*, 1956, **22**, (10), 1186–1189). [In Russian]. A versatile reflection-type ultrasonic flaw-detecting device working with a frequency of 160 kilohertz has been developed. Results are registered oscillographically. It has been used for testing metal tubes and bars.—S. K.

The Ultrasonic Testing of Heavy Forgings. G. Beckmann. (*Neue Hütte*, 1957, **2**, Feb.-Mar., 169–178). A description is given of the work on the ultrasonic testing of heavy forgings carried out by the Iron Research Institute in the East German Republic since 1954. Development of the method has now attained the level of other countries. Problems in the evaluation of test results are discussed.—L. J. L.

Ultrasonic Defectoscope Examination of Butt Welded Joints. A. K. Gurvich. (*Avtomaticheskaya Svarka*, 1956, (2), 68–75). [In Russian]. The ultrasonic defectoscope method developed by S. Ya. Sokolov reveals cavities, slag inclusions and cracks which are even beyond the scope of gamma-radiography. The effectiveness of the defectoscope method is two to four times higher than that of the X-ray method and three to five times higher than that of gamma-radiography.—R. S.

Demountable Artificial Defect for Evaluating Penetrants in Flaw Detection. R. W. Roberts and A. D. Dudbridge. (*Metallurgia*, 1957, **55**, June, 311–312).

Material Testing. R. A. Schamschula. (*Maschinenwelt Elektrotechnik*, 1956, **11**, Apr., 112–115). This article gives a brief outline of techniques used in non-destructive testing of metals, and deals with spark testing, radiographic examination, testing with gamma radiation, and magnetic and ultrasonic testing.—L. D. H.

Application of Non-Destructive Testing to Castings. C. W. H. du Toit. (*Engineer Foundryman*, 1957, **21**, Feb., 58–62; discussion, 62–63). The author discusses how the application of non-destructive testing methods can help to improve the quality of castings and gives examples of the use of radiography.—A. D. H.

Gamma Radiography in Steel Foundries. G. M. Michie. (*Engineering*, 1957, **184**, July 12, 46). Layout and protective measures are briefly considered.

Electron Diffraction Study of Flake Graphite Extracted from Molten Pig Iron. K. Niwa and G. Shimaoka. (*Trans. Amer. Inst. Min. Met. Eng.*, 1957, **209**; *J. Met.*, 1957, **9**, Apr., Section 2, 430–434). Electron-diffraction studies of single graphite flakes segregated from molten carbon-rich or silicon-rich iron at 1200°–1500° C are reported. The well-developed thick flakes are composed of one single crystal, the thin almost colourless flakes consisting of two or three single crystals. The orientations and stacking of atomic layers are discussed.

The Use of Radioactive Isotopes for Non-Destructive Testing of the Macrostructure of Materials in the Klement Gottwald Steelworks in Ostrava. O. Kožušník. (*Zváranie*, 1956, **5**, (5), 146–150). [In Czech]. Practical applications of isotopes, particularly ^{60}Co, are described.—P. F.

Determination of Thermal Conductivities of Metals by Measuring Transient Temperatures in Semi-Infinite Solids. S. T. Hsu. (*Trans. Amer. Soc. Mech. Eng.*, 1957, **79**, July, 1197–1203).

Thermal Expansion Coefficient, Rigidity Modulus and its Temperature Coefficient of the Alloys of Iron, Nickel, Cobalt and Chromium. I. Thermal Expansion and the Relation between Super-invar and Stainless Invar. H. Masumoto, H. Saito, and T. Kono. **II. Rigidity Modulus and its Temperature Coefficient, and the Relation between Co-elinvar and Elinvar.** H. Masumoto, H. Saito, and Y. Sugai. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Aug., 478–480; 481–484). [In Japanese]. I. In the quaternary Fe–Co–Ni–Cr system, there are two " valleys " of the expansion coefficient, one originating in the composition of Invar or Super-invar, and the other in that of stainless Invar. These come together at the composition 10% chromium; the nickel and cobalt in these alloys can be substituted for one another in almost linear relations. II. There are two portions of the system showing maximum positive values of the temperature coefficient of rigidity modulus, one corresponding with the composition of Elinvar and the other with that of Co-elinvar. These coincide at the composition of approximately 14% of nickel, 35% of cobalt, 10% of chromium, and the remainder iron; the nickel and cobalt in these alloys may be substituted for one another.—K. E. J.

Influence of Addition of Nickel on the Thermal Expansion, the Rigidity Modulus and its Temperature Coefficient of the Alloys of Cobalt, Iron and Vanadium. I. Alloys Containing 9·1% of Nickel. H. Masumoto, H. Saito, and S. Tetuka, **II. Alloys containing 20 and 30% of Nickel.** H. Masumoto. H. Saito, and K. Goto. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Aug., 484–486; 487–489). [In Japanese]. I. The effect of the Ni on the coeff. of thermal expansion and the temp. coeff. of the rigidity modulus is as if the Co has been replaced by the Ni. When Ni is added, the concentration range having the positive temp. coeff. of rigidity modulus becomes wider, and the slope of the latter to the concentration smaller. II. The effects of additions of 10% and 20% of Ni follow the same trends as those of the 9·1% addition.—K. E. J.

High Temperature Dilatometer. (*Canad. Mines Tech. Surveys, Phys. Met. Division*, (*July-Dec.* 1956), 1957, Jan. 15, 2–3). The device operates up to 1200° C.

Effect of Silicon and Aluminium Content on Mechanical and High-temperature Properties of Nodular Cast Iron. N. Shinohara. (*Nippon Kinzoku Gakkai-Si*, 1955, **19**, Sept., 525–528). [In Japanese]. The work was done with the object of obtaining irons of superior mechanical and high-temperature properties by promoting formation of spheroidal graphite in Si–Al irons. About 1% of Al promoted, but over 3% impeded, the spheroidisation of graphite. The growth and scaling of 3·8% Si, 5% Al nodular iron and 5·4% Si, 3% Al nodular iron were

almost inappreciably small, even at 900° C. Irons containing high proportions of Si and Al were not improved by Mg additions, but those containing low proportions were.—K. E. J.

Influence of Phosphorus and Silicon on the Properties of 13 Cr Stainless Steels. K. Kusaka. (*Tetsu to Hagane*, 1957, **43**, Mar., 327–328). [In Japanese]. For steels in the 0·08 C, 13·5 Cr and 0·17 C, 13·5 Cr series of varying compositions (given), the A_c and A_r points are listed, the effects of P and Si on the mechanical properties are illustrated, and changes in hardness through cold reduction (up to 70%) and annealing are shown.—K. E. J.

[5% Cr Steel as Heat-resisting Steel]. T. Kuno. (*Tetsu to Hagane*, 1957, **43**, Mar., 317–318). [In Japanese]. Mechanical properties at room temperature and 407°, 482°, 538° and 563° C, are given for four different 5% Cr Steels (full analyses given); they are continued on graphs up to 800° C.

Nickel Alloys with Special Properties. C. G. Smith. (*J. B'ham. Met. Soc.*, 1957, **37**, Sept., 553–565). The Ni–Fe alloys are reviewed with properties and uses. Low-expansion alloys, glass-to-metal seals, constant-modulus alloys, high-permeability alloys, permanent magnets, magnetostrictive compositions and magnetic compensation alloys and their applications are considered.

Studies on Nickel-Base Heat-Resisting Alloys. II. T. Hasegawa. (*Tetsu to Hagane*, 1957, **43**, Mar., 274–275). [In Japanese]. For M252 alloy, containing 52·13% Ni, 19·72% Cr, 10·20% Mo and 9·76% Co, the effects of fourteen different heat-treatments on the room-temperature tensile and creep-rupture properties are presented.—K. E. J.

Tool Steels. V. Materials for Machine Tool Service. L. F. Spencer. (*Steel Processing*, 1957, **43**, Sept., 511–516, 526). Properties and uses are tabulated and choice of steels for different purposes is discussed. Water and oil hardening steels are referred to and the heat treatment of high speed steels is reviewed with notes on tempering. Types T1, T2, T4, T5, T6, T8, T9, M2, M3, M10 and M35 are then separately considered and uses indicated.

The Solubility of Nitrogen in Iron–Chromium Alloys. E. T. Turkdogan and S. Ignatowicz. (*J.I.S.I.*, 1958, Mar., **188**, 242–247). [This issue].

The Fifth Anniversary of the Iron Research Institute at Henningsdorf near Berlin. E. Maurer. (*Abhandlungen der Deutschen Akademie der Wissenschaften zu Berlin*, 1956, 59–75). After reviewing the founding, lay-out and work carried out by the Institute, the author goes on to give some specific examples of problems dealt with. These include co-operative work for new standard specifications, consulting work for the production of turbine blades, high-temperature steels, manganese steel, nickel-free steel for railways and high-speed steels. (12 references.)—R. P.

METALLOGRAPHY

Replicas of the Structure of Metals with the Aid of Titanium. A. I. Rizol' and L. M. Utevskii. (*Zavodskaya Laboratoriya*, 1956, **22**, (5), 567–569). [In Russian]. The advantages of using Ti (99%) for making replicas of actual surfaces for electron-microscopical examination are considered and some results obtained with steel illustrated.—S. K.

The Agreement of Evaluations of the Carbide Lattice in Steel from Annealed and Hardened Polished Sections. N. K. Ipatov. (*Zavodskaya Laboratoriya*, 1956, **22**, (5), 563–564). [In Russian]. An account is given of a statistical treatment of parallel estimates of carbide-lattice estimates on annealed and on hardened specimens of the same heats of bearing steel. Higher values are probable with the hardened specimens.

On the New Oxidizing Method of Observing Austenite Grains. H. Hirotani. (*Tetsu to Hagane*, 1957, **43**, Mar., 344–345). [In Japanese]. A thermal etching method involving borax was used on various plain-carbon, low-alloy and high-alloy steels, and shown to be more efficient than the normal method. It was used to study the effect of nitrogen on the austenitic grain size.—K. E. J.

Method of Measuring Austenitic Grain. I. E. Kontorovich and B. M. Voshedchenko. (*Zavodskaya Laboratoriya*, 1956, **22**, (8), 954–955). [In Russian]. The etching of austenite grain-boundaries of three heat-treated steels with a picric-acid–ether–water reagent is briefly described, illustrated and discussed.—S. K.

Veining in Ferrite. A. Hultgren, Å. Josefsson, E. Kula, and G. Lagerberg. (*J.I.S.I.*, 1958, Mar., **188**, 247–261). [This issue].

Visualization of the Structure of Iron–Alloy Surface Layers Not Amenable to Etching. B. A. Krasyuk. (*Zavodskaya Laboratoriya*, 1956, **22**, (5), 556–558). [In Russian]. The problem of structure visualization for steel surfaces which have been diffusion-annealed or have been subjected to some special (e.g., spark) treatment rendering them unetchable by normal methods is discussed. Experiments are described in which electrochemical potential determinations were made on a carbon steel Cr-alloyed on its surface and on a spark-treated tool steel, the unetchable surface-layers being 0·1 and 0·08 mm thick, respectively. It was found that the protective action of the base metal towards the surface could be neutralized by contacting the latter with a metal, such as silver, having a higher electrochemical potential.

Study on Carbides in Steel by Electrolytic Isolation. I. Electrolytic Isolation of Carbides from Carbon Steel by the Method of Klinger and Koch. K. Tasaka. (*Tetsu to Hagane*, 1957, **43**, Mar., 408–409). [In Japanese]. For steels between 0·16–0·99% C (compositions given), optimum current densities are established for isolation of cementite from annealed and spheroidized specimens.—K. E. J.

Study on Carbides in Practical Special Steels by Electrolytic Isolation. V. On Carbides in High-C High Cr–W Die Steel. T. Sato. **VI. On Carbides in Hot Working Die Steel.** T. Nishizawa. (*Tetsu to Hagane*, 1957, **43**, Mar., 405–406; 406–408). [In Japanese].—K. E. J.

Studies of Ball Bearing Steels. VIII. Behaviour of the Carbide in Ball Bearing Steels by Electrolytic Isolation. M. Ueno and H. Nakajima. (*J. Mech. Lab.*, 1957, **3**, (1), 33–38). Cementite was isolated from SKF bearing and basic electric furnace steels and analysed so as to obtain the composition of the matrix. Annealed ball-bearing steels contain about 15% Fe_3C with a higher concentration of Cr, Mn and V, but less Ni than the matrix, these concentrations change somewhat with austenitizing temperature. Steels hardened at 820–840° C have 7–9% cementite and in the two steels show differences in Cr and Mn contents.

Studies on Ball Bearing Steels. X. Surface Potential Change of Ball-Bearing Steels in Electrolytic Isolation of Carbides. H. Nakashima. (*Tetsu to Hagane*, 1957, **43**, Mar., 409–411). [In Japanese].—K. E. J.

Regarding the Theoretical Basis for Determining the Carbide Phase in Carbon Steels. Yu. Klyachko and O. D. Larina. (*Zavodskaya Laboratoriya*, 1956, **22**, (12), 1409–1415). [In Russian]. The phase analysis makes use of the difference in the chemical potentials or in the free energies of the dispersed phases. Consequently the theoretical basis and the evolution of electrochemical methods of phase analysis resides in the investigation of polarization phenomena. These are described in detail with various electrolytes and varying temperatures and current densities. The results, in the opinion of the authors, indicate the way to a solution of the problems of electrochemical phase analysis.—L. H.

Evidence from Autoradiography for Segregation of Traces of Sulphur in the Intergranular Boundaries of Annealed α-Iron. C. Leymonie, P. Coulomb and P. Lacombe. (*Compt. Rend.*, 1957, **245**, Sept. 2, 931–934). Support is found for the hypothesis that abnormal segregation occurs above the limiting solubility temperature in grain boundaries.

Study on the Rapid Photometric Determination of Various Components of Non-Metallic Inclusions in Steel. T. Ikegami. (*Tetsu to Hagane*, 1957, **43**, Mar., 389). [In Japanese]. Details are given of methods for SiO_2, Al_2O_3, FeO, Cr_2O_3, TiO_2 and MnO.—K. E. J.

The Recrystallization Texture of Low Carbon-Steels after Cold Deformation. K. V. Grigorov and G. P. Blokhin. (*Fizika Metallov i Metallovedenie*, 1957, **4**, (2), 331–338). [In Russian]. This recrystallization structure depends on the stages of deformation and on its own character. Two textures exist simultaneously. The first type (001) [100] and (100) [001] clearly occurs in weak deformations, but with the increase in the amount of deformation the texture grows in the order (001) [100] ~ 15°; and (112) [110] ~ 15°. Annealing temperatures between 700° and 820° do not influence the character of the structure or the stages of its formation.—L. H.

Determination of the Linear Rate of Growth and Spread of the Disperse Phase. A. G. Spektor. (*Zavodskaya Labora-*

toriya, 1956, **22**, (12), 1455–1458). [In Russian]. An investigation is described for a 0·4% C steel in which the rate of growth of the grains was determined during tempering at 630° C. The results show that in the cementite phase there exists a critical grain diameter which behaves 'neutrally' i.e. that on either side of it the rate of growth of the grains differs in speed and direction.—L. H.

The Recrystallisation Texture of Cold Drawn Steels. Ts. N. Rafalovich. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (2), 326–331). [In Russian]. During the recrystallization treatment of steels the newly appearing grains possess an orientation identical with the deformed matrix within which they appeared. The main orientation of the crystals in the deformed texture of the matrix is maintained in the recrystallization texture, the degree, only, of the scattering around it is different.—L. H.

Comparison of Dislocation Densities of Primary and Secondary Recrystallization Grains of Si–Fe. C. G. Dunn and E. F. Koch. (*Acta Met.*, 1957, **5**, Oct., 548–554).

On the Polymorphism Theory of Steel. V. N. Svechnikov and A. G. Lesnik. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (1), 87–96). [In Russian]. Regarding, as a first approximation the solid solutions of iron with the limiting field elements as ideal solutions, the authors show that the phase equilibrium curve $\gamma/\gamma + \alpha$ can be quite well approximated in many cases by the 2nd order curves. The equation $\gamma/\gamma + \alpha$ and $\gamma + \alpha/\alpha$ was obtained from the phase equilibrium of the system Fe–Cr. The theoretical and experimental graphs agree satisfactorily.—L. H.

The Isothermal Transformation of Spheroidized Pearlite to Austenite. A. A. Golestaneh, C. Handford, and G. Bullock. (*J.I.S.I.*, 1958, Mar., **188**, 237–241). [This issue].

The Influence of Irradiation by Neutrons on the Martensite Transformation. A. I. Zakharov and O. P. Maksimova. (*Doklady Akad. Nauk, S.S.S.R.*, 1957, **114**, (6), 1195–1198). [In Russian]. Various types of steel and iron alloys were subjected to neutron bombardment after annealing at various temperatures. It was found that the effects of activation or retardation of the austenite–martensite transformation by neutron bombardment, were rather complex and depended not only on the temperature, but also on the nature of the material.—S. I. T.

Heterogeneity of Martensite with Regard to Carbon Content in Steel Specimens After High Frequency Hardening. I. N. Kidin. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (2), 299–305). [In Russian]. The investigations were devised so as to observe the behaviour of martensite with regard to the carbon in steels heated very rapidly by high frequency currents (up to 4500°/sec) with the respective graphs photometrically recorded. The results seem to show the growth of heterogeneity in the solid solution up to 840° and its diminution beyond that temperature. The graph maxima describe the conditions for the appearance of austenite crystals highly saturated with carbon. The increase in temperature reveals an intensive migration of carbon atoms in those sections where the fall in concentration is greatest.

Relation Between the Carbon Content of Martensite and the Heating Parameters of Steel Specimens in High Frequency Hardening. I. N. Kidin. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (2), 306–308). [In Russian]. The investigations here described justify the assertion that the great hardness after high-frequency hardening does not depend only on the total carbon content of the steel, but also on the carbon distribution in the martensite crystals. The variations of the hardness with the increase in the hardening temperature, is, among other factors, due to the distribution of carbon during the heating.—L. H.

On the Reduction of the Temperature at the Start of the Martensite Transformation During the Partial Dissociation of Austenite in the Bainite Region. V. I. Arkharov, V. P. Sklyuev and V. O. Esin. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (1), 62–65). [In Russian]. To investigate the critical point of diffusion of austenite, use was made of the Zeiss differential optical dilatometer and the results justify the conclusion that the carbon concentrations in austenite regions at various stages of cooling, deviate less from the average carbon content than the concentrations in the small nuclei initiating each successive stage in the process.—L. H.

Studies of the Relation between Martensite and the Non-acicular Constituent Occurring with Martensite in Fe–Mn–C Alloys by the Bitter-Akulov Method. A. Mašin, V. Havel and D. Tlusta. (*Bull. Acad. Polonaise Sci.*, 1957, **5**, (3), 181–184). [In German]. The phase appears in 5% Mn, 1% C alloy after quenching from 1150° C and cooling below 0°. A magnetic method was used for the investigation and etched sections are shown. The constituent is ferromagnetic like martensite and it is concluded that it is martensite in a different physical condition.

The Formation of Martensites in Fe–Mn–C Alloys. A. Mašin. (*Bull. Acad. Polonaise Sci.*, 1957, **5**, (3), 185–190). The structures presented by martensite in steel with 1% C and 5% Mn are discussed. Chilled specimens were cooled successively to − 10°, − 20°, − 40°, − 55°, − 77° and − 78° C and the microstructure examined and the changes in shape and dimensions are described and hypotheses to account for these are advanced.

Classification of Ternary Metallic Phases. B. K. Vul'f. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (1), 97–109). [In Russian]. The author has worked out a classification of ternary metallic phases on the principle of comparison of the extent and character of the arrangement of the homogenous regions on ternary diagrams. He points to the contribution which his classification makes to the elucidation of the Periodic Law within the limits of each classificational type. The proof is also supplied that in a series of cases the resulting ternary metallic phase is formed by the interchange of analogous elements in the Periodic Table. A comparison is also made between the ternary metallic phase and the previously available systems of classification of binary phases. (189 references).—L. H.

Heat Treatment of Dispersed System Magnet Steels. VII. On Some Experimental Results with X-Ray Diffraction in the Fe–Ni–Al and Fe–Ni–Al–Co Systems. M. Chubachi and H. Ueshima. (*Sumitomo Metals*, 1957, **9**, Apr., 118–123). The appearance of the gamma-phase was observed and considered to be due to decrease of Al at the surface during heat treatment and a superlattice, α'-phase, was present showing forbidden α-phase planes. High coercive force appears to be due to changes in this α-phase.

Influence of Additions of Nickel on the Thermal Expansion, the Rigidity Modulus and its Temperature Coefficient of the Alloys of Cobalt, Iron and Vanadium. H. Masumoto, H. Saito and K. Goto. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1957, **9**, June, 159–169). The range 10–50° is covered for alloys with 9·1, 20 and 30% Ni. Nickel replaces cobalt so that the alloy with lowest thermal expansion and largest temperature coefficient of rigidity shift to lower Co content on addition of Ni, also the maximum becomes flatter and the change of coefficient with concentration becomes smaller.

Quantitative Phase Analysis of Kovar Alloys by X-Ray Diffraction. A. Spirkov. (*Hutnické Listy*, 1957, **12**, (7), 609–614). The volumetric method of Fletcher and Cohen is described. In the case of Kovar the values of G_α and G_γ have to be measured experimentally whereas in the case of steel they can be calculated as the characteristic temperature of both phases is known. The values for the temperature function B(T) are shown to be different for each phase.

CORROSION

On Below-Dew-Point Type Corrosion. I. S. Jo. (*Tetsu to Hagane*, 1957, **43**, Mar., 314–316). [In Japanese].—K. E. J.

The Influence of the Alloy Components on Heat Resisting Alloys and on the Binding Force during the Phase of Oxidation of the Lattice in the Scale. The Influence of Chromium on the Binding Force in Haematite. V. I. Arkharov and B. S. Borisov. (*Fizika Metallov i Metallovedenie*, 1956, **3**, (3), 471–476). It appears from the data obtained that the participation of Cr in the composition of haematite increases the binding force within its lattice, and that this increase grows with the Cr content in the haematite.—L. H.

Influence of Light on the Corrosion of Iron by (Aqueous) Solutions. B. Lovrečeк and E. Korkut. (*Werkstoffe Korrosion*, 1957, **8**, May, 277–280). Experiments are reported on the corrosion of iron in NaOH and Na_2HPO_4 solutions. The losses in weight of specimens exposed to the light were about 20% greater than those of similar specimens kept in the dark. There were, however, no significant differences between corresponding potential/time curves under the two sets of conditions.—J. C. H.

Stray-Current Corrosion of Underground Distribution Systems. D. W. Barron. (*J. Amer. Water Works Assoc.*, 1957, **49**, July, 881–886).

Corrosion of Steel Structures in Water. P. Dutko. (*Oceliarsky Sbornik*-1954, 1955, 391–400). [In Slovak]. The nature and methods of prevention of corrosion of underwater steel structures are discussed with special reference to the Czechoslovak Standard Specification CSN 730911 relating to the protecting of steel structures in hydraulic engineering. Apart from the more common methods of protection special shapes of sections and Cu- and Cr additions can be utilized to minimize attack in water. Welded low-P profiles are particularly suitable.—P. F.

Recent Metallurgical Problems in Marine Engineering. B. Todd. (*Trans. Inst. Marine Eng.*, 1957, **69**, Aug. 320–325). Corrosion of shaft liners and brittle fracture of wrought iron are considered as well as non-ferrous topics; also the cracking of spheroidal graphite iron pistons.

Studies on Ferrous Materials for Nuclear Reactors. I. Static and Stress Corrosion in Uranyl Solutions. M. Hasegawa. (*Tetsu to Hagane*, 1957, **43**, Mar., 316–317). [In Japanese]. An apparatus is described in which specimens can be given static corrosion or stress corrosion tests at various temperatures in corrosive uranyl solutions.—K. E. J.

B.I.S.R.A. Corrosion Research. Part I. Corrosion of Bare Ferrous Metals. J. C. Hudson. (*Corrosion Tech.*, 1957, **4**, Sept., 320–324. A summary of results during the 30 years of the existence of the Corrosion Committee is given. Atmospheric corrosion, with the use of low-alloy steels, corrosion by fresh and salt waters and corrosion by soil are considered, and also protection by paint and metallic coatings.

The Behaviour of Materials with Respect to Hydrogen Fluoride and Hydrofluoric Acid. E. Lingnau. (*Werkstoffe Korrosion*, 1957, **8**, Apr., 216–233). A description is given of the action of gaseous and aqueous hydrofluoric acid on a number of metals and non-metallic materials. Carbon steels are resistant to HF in concentrations greater than 75% up to 65° C, but steel containers for the transport of aqueous HF of $>$ 60% should be previously passivated. Mild steel is resistant to gaseous HF, but cast iron is attacked by the acid at all concentrations.—L. D. H.

Method of Testing Steam Boiler and Furnace Steels. V. A. Kislik and V. N. Tkachev. (*Zavodskaya Laboratoriya*, 1956, **22**, (12), 1482–1486). [In Russian]. Four series of tests are described in a specially constructed boiler, in order to elucidate the corrosion processes and to discover means for their elimination. The chemical composition of the boiler water has been qualitatively changed for three series of experiments dealing with the corroding effects on the protective oxide film of the metals, and a fourth series of tests dealt with the friction effects of overlapping surfaces. Some definite conclusions emerged from the findings.—L. H.

Modern Views on the Passivity of Stainless Steel. M. G. Fontana and W. P. McKinnell, jun. (*Werkstoffe Korrosion*, 1957, **8**, May, 249–254). After a short review of views on passivity since the time of Faraday (1844), the authors describe recent work at Columbia University in which the adsorption of O_2 at low pressures on stainless steel surfaces was studied. The results confirm the oxide film theory of passivation and suggest that this film consists of amorphous hydroxides.—J. C. H.

Intergranular Corrosion Test of Austenitic Stainless Steels with Acidic Copper Sulphate Solution Containing Copper. N. Ito. (*Tetsu to Hagane*, 1957, **43**, Mar., 313–314). [In Japanese]. Results are given for six steels, of the A.I.S.I. 304, 304L, 316 and 316L types (analyses given), showing the increasing degree of electrical resistance caused by intergranular corrosion, for periods from 5 to 72 h. The effects of Cu on the redox potential of $CuSO_4$–H_2SO_4 solution and on the electrode potential of type 316L steel are shown.

Corrosion Research Laboratories—13. Ministry of Supply Tropical Testing Establishment, Port Harcourt, Nigeria. A. A. J. Bain. (*Corrosion Tech.*, 1957, **4**, Oct., 357–360). Exposure tests are discussed, and effects of light intensity and biological factors.

Protection of Steel Piles in Norway. I. T. Rosenqvist. (*Corrosion Tech.*, 1957, **4**, Aug). Buildings on steel piles are becoming common in Norway and the metal is usually left unprotected. Suitable sections and cathodic protection in unfavourable soils are suggested.

U.S. Tankers Freed From Corrosion. J. Grindrod. (*Corrosion Tech.*, 1957, **4**, Aug., 278–280). Methods of cathodic protection of mills, tanks, and pipes on two ships chartered by the Dow Chemical Co. are described. Mg anodes were fitted before launching to one vessel. Relative maintenance costs are tabulated.

Cathodic Protection in Israel. Part I. On the Land. D. Spector. (*Corrosion Tech.*, 1957, **4**, Aug., 265–268). **Part II. On Sea and In the Home.** (Sept., 306–309). Irrigation works are described in the first part and fishing vessels and electric hot water systems in the second.

Zinc Anodes for Cathodic Protection. J. H. Morgan. (*Corrosion Tech.*, 1957, **4**, Aug., 272–274). Suitable alloying of zinc for electrodes and the shape, size and location of the anodes are discussed. Protection of Al and Pb with Zn is also mentioned. Suitable potentials are noted.

Instruments for Cathodic Protection. (*Corrosion Tech.*, 1957, **4**, Aug., 269–271). Design of half-cells for technical purposes and means for potential measurement, resistivity and its determination and the examination of coatings for buried pipe-lines and testing for continuity are briefly reviewed.

Stopping Corrosion with Rust Preventive Compounds. R. P. Mills. (*Iron Steel Eng.*, 1957, **34**, Sept., 96–98, Discussion 98–100). An elementary account of corrosion processes and the petroleum compounds used to prevent them, with notes on testing. Some particular cases are mentioned in discussion.

ANALYSIS

Colorimetric Determination of Iron with *aa'*-Dipyridyl. M. Grat-Cabanac. (*Anal. Chim. Acta*, 1957, **17**, Sept., 348). [In French]. Determination of Fe in a Pt alloy is described, small amounts of Zn do not interfere.

Determination of Iron in Bimetallic (Copper-Steel) Ingots and Wires by a Thermoelectric Method. P. D. Korzh. (*Zavodskaya Laboratoriya*, 1956, **22**, (9), 1076–1078). [In Russian]. Based on studies of the dependence of the e.m.f. of Cu–Fe alloy/Cu/Fe couples on temperature and also on alloy iron-content, the Fe content of the Cu of Cu-steel materials has been determined.—S. K.

A Vacuum Fusion Apparatus for Gas Analysis. P. D. Blake. (*J.I.S.I.*, 1958, Mar., **188**, 261–264). [This issue].

Rapid Photometric Determination of Phosphorus in Steel. S. Wakamatsu. (*Tetsu to Hagane*, 1957, **43**, Mar., 389–390). [In Japanese].—K. E. J.

Photocolorimetric Determination of Phosphorus in the Presence of Tungsten, Titanium and Niobium. E. I. Nikitina. (*Zavodskaya Laboratoriya*, 1956, **22**, (9), 1027–1030). [In Russian]. Conditions for the photocolorimetric determination of P in steels and other alloys containing W, Ti and Nb with the combination of the interfering components into complexes have been investigated.—S. K.

Non-Destructive Method for Determining Phosphorus in Ferrous Alloys. N. A. Tananaev and I. A. Korobova. (*Zavodskaya Laboratoriya*, 1956, **22**, (8), 916–917). [In Russian]. In the non-destructive method described, which can be used to determine P in low-alloy steels and irons, 1 : 1 nitric acid is used as the solvent. The reaction for PO_4^{---} is the formation of the orange-yellow complex $[P(Mo_2O_{10})_2(V_2O_6)_2]^{---}$ formed with a previously mixed molybdate-vanadate reagent.—S. K.

New Rapid Method for the Determination of Gaseous Contents in Iron and Steel. III. On the Evaporated Film of Metals. N. Yoneda. (*Tetsu to Hagane*, 1957, **43**, Mar., 395–396). [In Japanese].—K. E. J.

Photocolorimetric Determination of Vanadium and Titanium. E. F. Pen'kova, A. V. Gladkova and T. V. Novikova. (*Zavodskaya Laboratoriya*, 1956, **22**, (8), 918). [In Russian]. For the determination of V in steels containing Mo, Mn and Cr a photocolorimetric method with H_2O_2 has been developed and tested. For Ti in steel a variant of the chromotropic acid method was found to be satisfactory.—S. K.

Use of Ascorbic Acid in Amperometric Titration. I. Determination of V and Cl in Presence of Other Elements. Z. A. Gallai, V. G. Tiptsova and M. Peshkova. (*Zhur. Anal. Khim.*, 1957, **12**, (4), 469–475). V is determined (in presence of Ni, Mn, Zn, Al, Cr, Ti, Fe, Mo and W); also quadrivalent Ce in pure salts, and ferric iron.

Photometric Determination of Boron in Steel. I. U. Martynchenko and A. M. Bondarenko. (*Zhur. Anal. Khim.*, 1957, **12**, (4), 495–498). B can be determined with carmine and by adding phosphoric acid, separation of heavy metals is rendered unnecessary.

Determination of Magnesium and Iron in Titanium and its Alloys. L. M. Budanova, L. A. Nenasheva and T. V. Matrosova. (*Zavodskaya Laboratoriya*, 1956, **22**, (12), 1419–1421). [In Russian]. A colorimetric method for the determination of Mg impurities contained in Ti and its alloys is described, which allows estimation at very low concentrations. A combined chemical and colorimetric method of analysis which gives precise results for the estimation of iron impurities in Ti and its alloys is also given.—L. H.

Photometric Determination of Tin in Iron and Steel. III. Photometric Determination of Tin with Sodium Diethyldithiocarbamate. H. Goto and Y. Kakita. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1957, **9A**, (4), 253–261).

Determination of Silica in Tungsten Alloys and Ferro-Tungsten by the Phosphoric-Sulphuric Acid Method. M. V. Babaev. (*Zavodskaya Laboratorya*, 1956, **22**, (11), 1294–1296). [In Russian]. This method of analysis proposed in 1950 by Penkova and Yakovlev, has been subjected to a searching investigation and has been found unreliable. The interaction between silicic and phosphoric acids which has been demonstrated in the tests here described, had not been taken into account by the original proposers of this method. This interaction proceeds by the formation of a meta-acid which in turn reacts with the silicic acid to form $SiO_2.P_2O_5$, which can be considered as silicon metaphosphate $(SiO)(PO_3)_3$. It is this reaction which vitiates the results of analyses obtained.—L. H.

Study of $K\beta_5$-Bands of the X-Ray Spectrum of Iron and Chromium at Magnetic Transformation of Alloys of the System Fe–Cr. V. A. Kazantsev. (*Doklady Akad. Nauk, S.S.S.R.*, 1957, **114**, (1), 86–87). [In Russian]. Three alloys of Fe–Cr have been studied by means of X-rays at different temperatures. It is concluded that the transformation of alloys into a paramagnetic state, occurring under the influence of temperature and variable content of chromium, is accompanied by an increase of the energy of $K\beta_5$ photons of Fe and its decrease in Cr.—S. I. T.

INDUSTRIAL USES AND APPLICATIONS

Miscellaneous Uses of Manganese Steel in Metal Mining. (*Edgar Allen News*, 1957, **36**, Sept., 202–205). An illustrated review.

On Pipes for Low Temperature Service. J. Omori, E. Miyoshi, K. Kawano and H. Maruoka. (*Sumitomo Metals*, 1957, **9**, Apr., 98–117). Steels for low temperature use in petroleum synthetic industries are studied with respect to heat-treatment, cold working and welding. Al-killed low-C steel and 3·5% Ni steels are suitable at — 46° C and — 101° C respectively.

The Choice of Steam Pressures for Power Plant Utilising Ferritic Steel Tubes. H. Haferkamp. (*BWK*, 1957, **9**, Aug., 369–370). By means of thermodynamic calculations the optimum operating conditions were evaluated. With the steel 10CrMo910 the optimum was found to be 175 atm, with the steel 13CrMo44 the corresponding pressure was 200 atm. The steam temperature at the latter pressure is 520° C

Chromium–Molybdenum and Chromium–Molybdenum–Vanadium Steels for Power Plant and Refinery Service up to 1100° F. G. V. Smith. (*Trans. Amer. Soc. Mech. Eng. Preprint*, 56 A 214, 1956, pp. 32). A review from the literature with 76 references. Conclusions on strength and creep resistance are given.

Investigation on Super Heat-Resisting Alloys for Gas Turbines. VI. Effect of Fe and N_2 upon the Age-hardening of LCN-155 Alloy. N. Yukawa. (*Tetsu to Hagane*, 1957, **43**, Mar., 299–301). [In Japanese].—K. E. J.

Iron and Steel in Nuclear Power Plants. R. Kiessling. (*Jernkontorets Ann.*, 1957, **141**, (9), 483–506). [In Swedish].

New Materials, Including Aluminium, Steel and Plastic, Challenging Lead as Cable Sheathing. C. C. Shackford. (*Metals*, 1957, **28**, July, 9–10). The use of mild steel strip 0·12–0·16 in. thick wrapped over copper and welded is described.

HISTORICAL

How the First Cast Iron was Obtained. J. Piaskowski. (*Przeglad Odlewnictwa*, 1956, **6**, Nov., 335–340). [In Polish]. The author gives an historical survey of the earliest methods of casting and their development, with diagrams which are dated from the early middle ages up to XVIII century.

Patent "Pigs" of the 'Eighteen-Thirties. T. R. Harris. (*Foundry Trade J.*, 1957, **103**, Sept. 19, 349–350). An account of George Crane's work on the smelting of ironstone with anthracite. The information is derived from surviving letters.

Statistics of the Scottish Pig Iron Trade 1830 to 1865. R. H. Campbell. (*J. West Scotland Iron Steel Inst.*, 1956–7, **64**, 282–289). Number of furnaces erected and in blast, production, consumption, stocks, shipments and prices have been compiled.

Augusto Vanzetti—Italian Steel Foundry Pioneer. B. Boni. (*Fonderia Ital.*, 1957, **6**, Aug., 319–326; *Met. Ital.*, 1957, **40**, July, 531–538). [In Italian]. An account is given of the life and work of Augusto Vanzetti, the first technical director of the Terni Steelworks. He was responsible for introducing the manufacture of large steel castings into Italy. Hitherto these had been made only in England and Germany. Born in 1846, he died in 1894. The great development of the Terni company is in large part due to his efforts.—M. D. J. B.

ECONOMICS AND STATISTICS

The National Income and Steel Production. K. Andrysik. (*Hutnik*, 1957, **24**, Jan., 18–22). [In Polish]. The relationship between the national income and steel production of U.S.A., France, U.S.S.R., and Poland are tabulated. Future estimated figures for increase in production according to increase of national income are given, which are: for 1955—$4{\cdot}45 \times 10^6$ t, 1960—7×10^6 t, which mean an increase of 60% during the 5-year plan.

Problems of Financing New Steel Capacity. W. B. Boyer. (*Iron Steel Eng.*, 1957, **34**, July, 75–76). The problems facing financing of modern steel plant are discussed briefly and the need for companies to increase there earning capacity for self-financing is stressed.—M. D. J. B.

The Iron and Steel Industry in Europe. O.E.E.C. Iron and Steel Committee. (*O.E.E.C. Document*, IS(57)3, 1957, Aug., pp. 92). A review of production price trends, and statistics of trade in ore, scrap, coke, iron and ferro-alloys and finished products, with tables of raw material supplies, energy, investments, manpower, productivity and costs.

Problems in Improving the Economics of Iron and Steel Works. L. I. Spivakovskii. (*Stal'*, 1957, (9), 843–846).

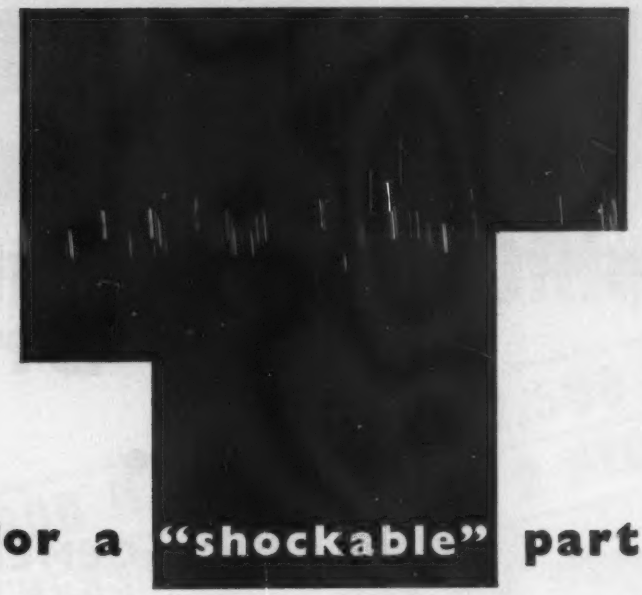
Analysis of the Deliberately Planned Losses in Some of the Heavy Industry Plants and of the Possibilities of Putting These Undertakings on an Economic Basis. C. Marcu. (*Met. Constr. Masini*, 1956, **8**, (6), 62–66). In the 1956 State plan, some large metallurgical installations are scheduled to work at a loss either because their costs of production are too high or because the selling prices have to be kept low. The author examines the situation created from all points of view and has some suggestions for rationalization and increase in productivity.—L. H.

MISCELLANEOUS

Endurance of Heat by Workers in the Metal Industry. W. Schweisheimer. (*Indust. Heating*, 1957, **24**, Aug., 1536, 1538, 1540). Al-coated protective clothing, studies of hot environments, effects on workers and working efficiencies are briefly described.

The Prevention of Gassing Accidents in Iron and Steel Works. I. S. Scott-Maxwell, D. Bruce-Gardner, J. W. Houghton, W. E. Scott and H. C. Stephenson. (*British Iron and Steel Federation Publication*, [1957], pp. 27). General practice and a special discussion of CO are considered.

Vol. 185, 1957, Jan., p. 141, col. 1, line 36. For "Bastier" read "Bastien." Line 51. For "Bruan" read "Braun." Feb. p. 273, col. 2, line 53. For "D. M. Petty" read "D. W. Petty." March. p. 430, col. 1, line 45. For "Grandall" read "Crandall." April, p. 562, col. 2, line 31. For "Healey" read "Healy." p. 563, col. 1, line 4. For "Kasmider" read "Kosmider." p. 587, col. 2, line 36. For "Weine" read "Wiene."



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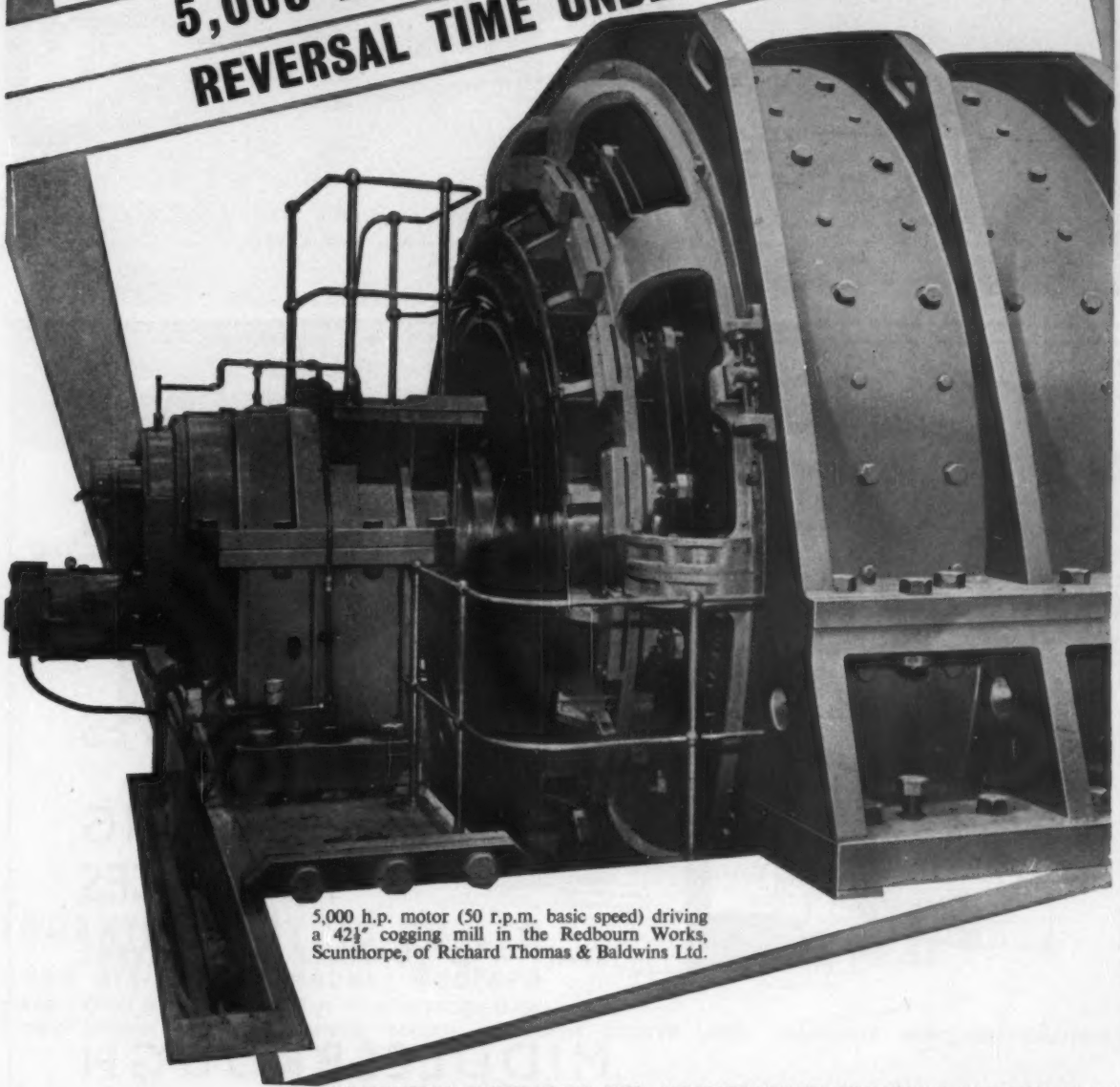

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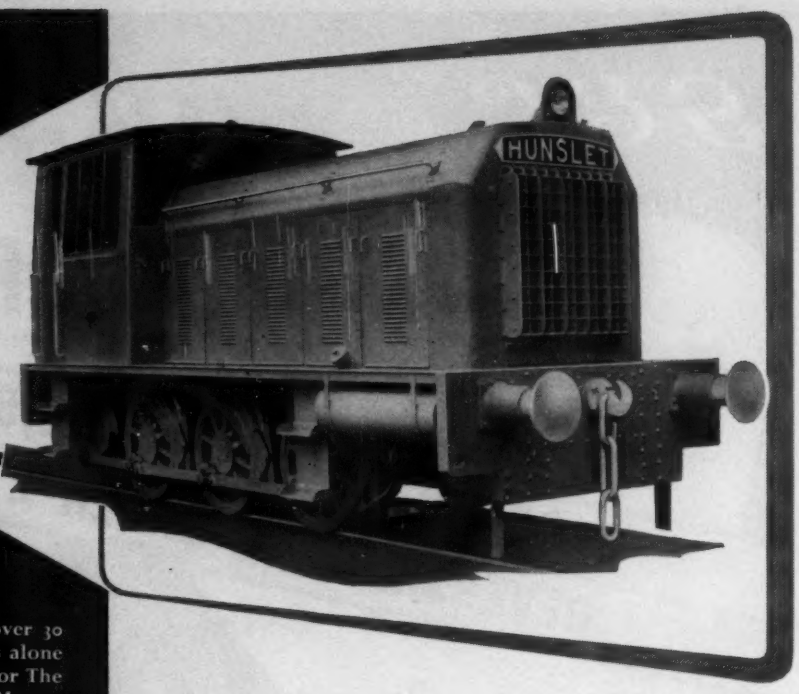

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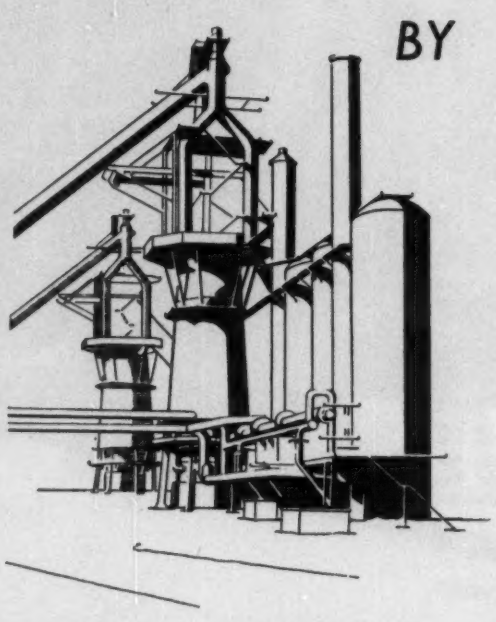

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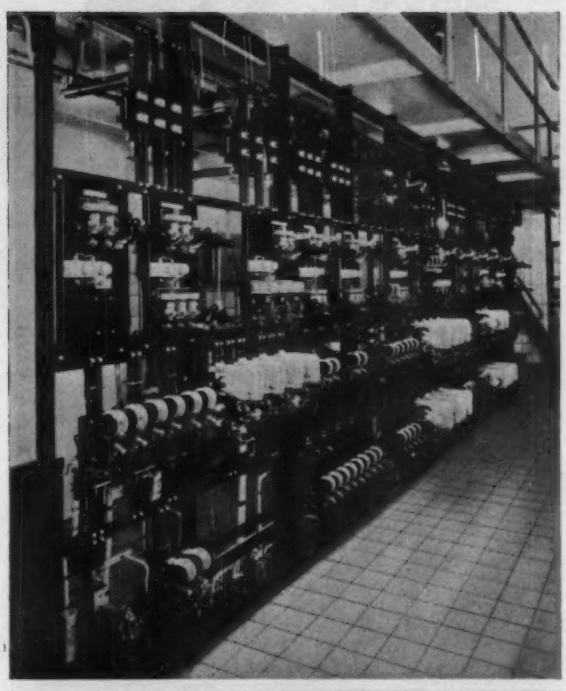

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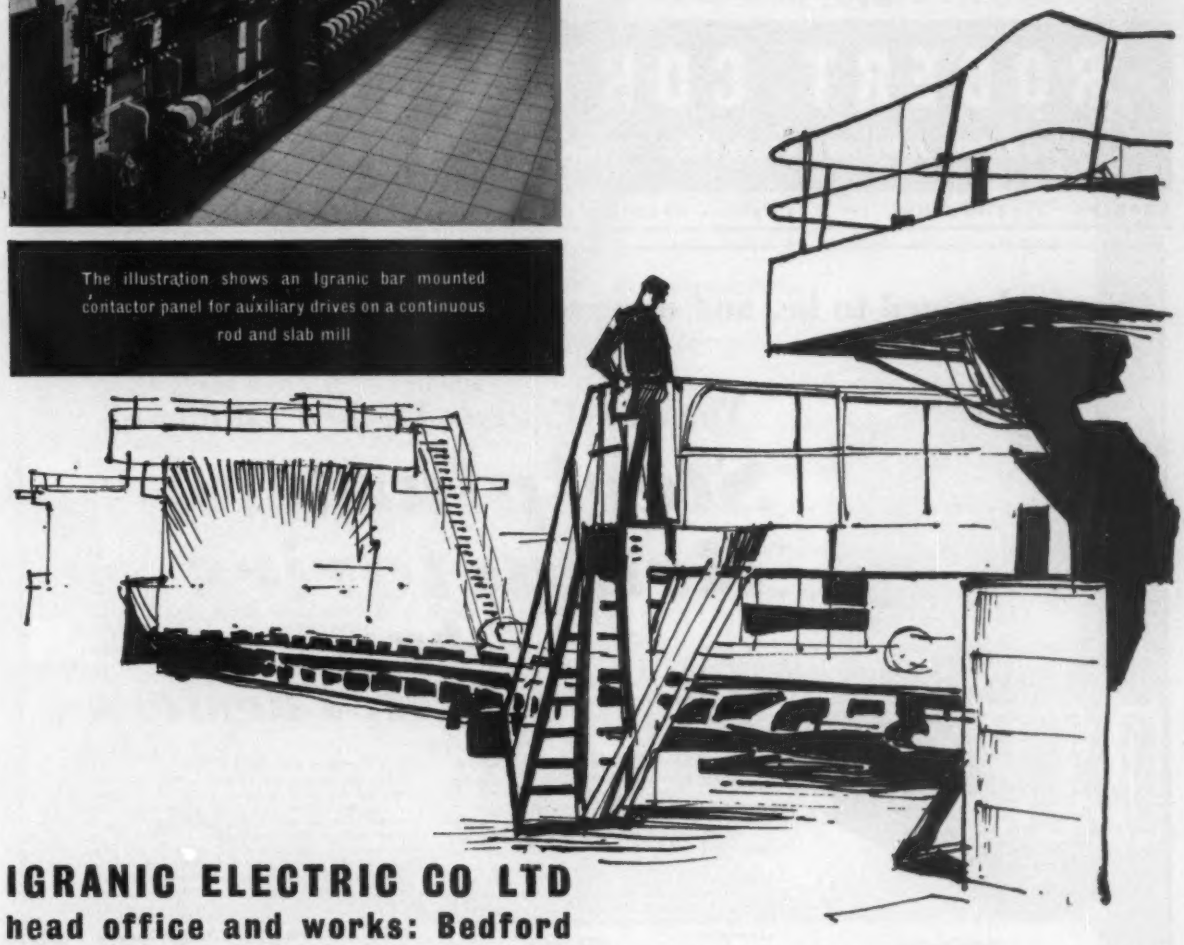

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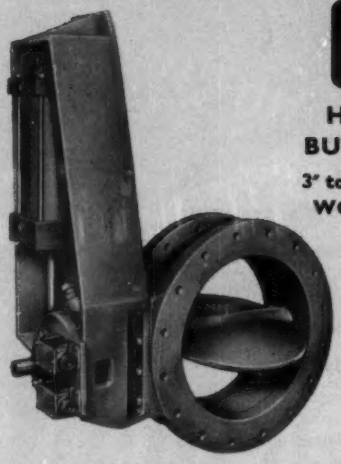

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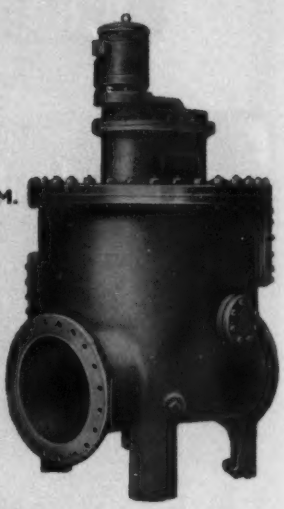

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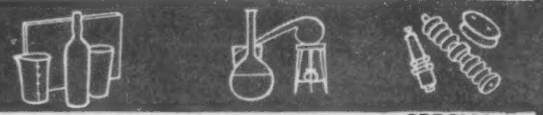

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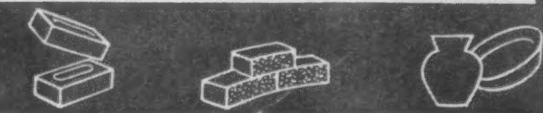

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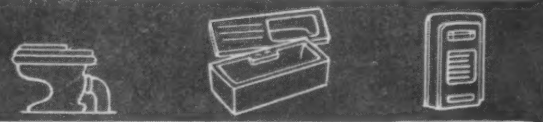

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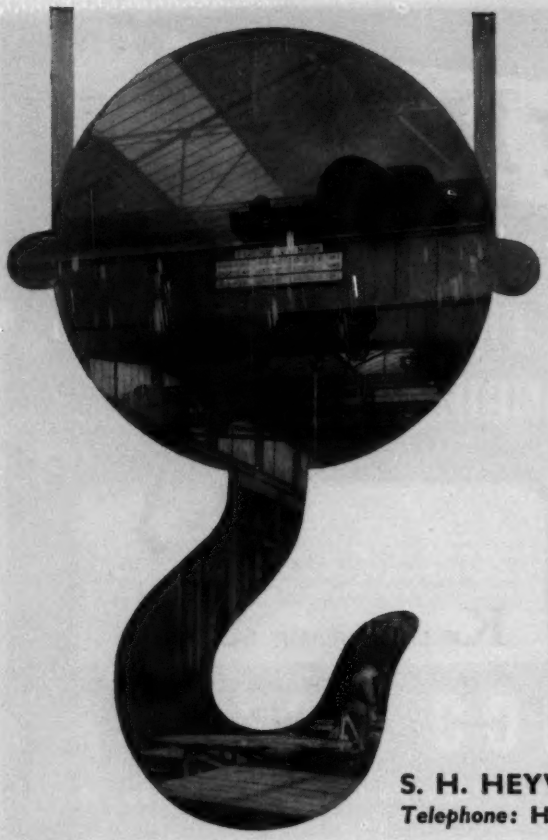

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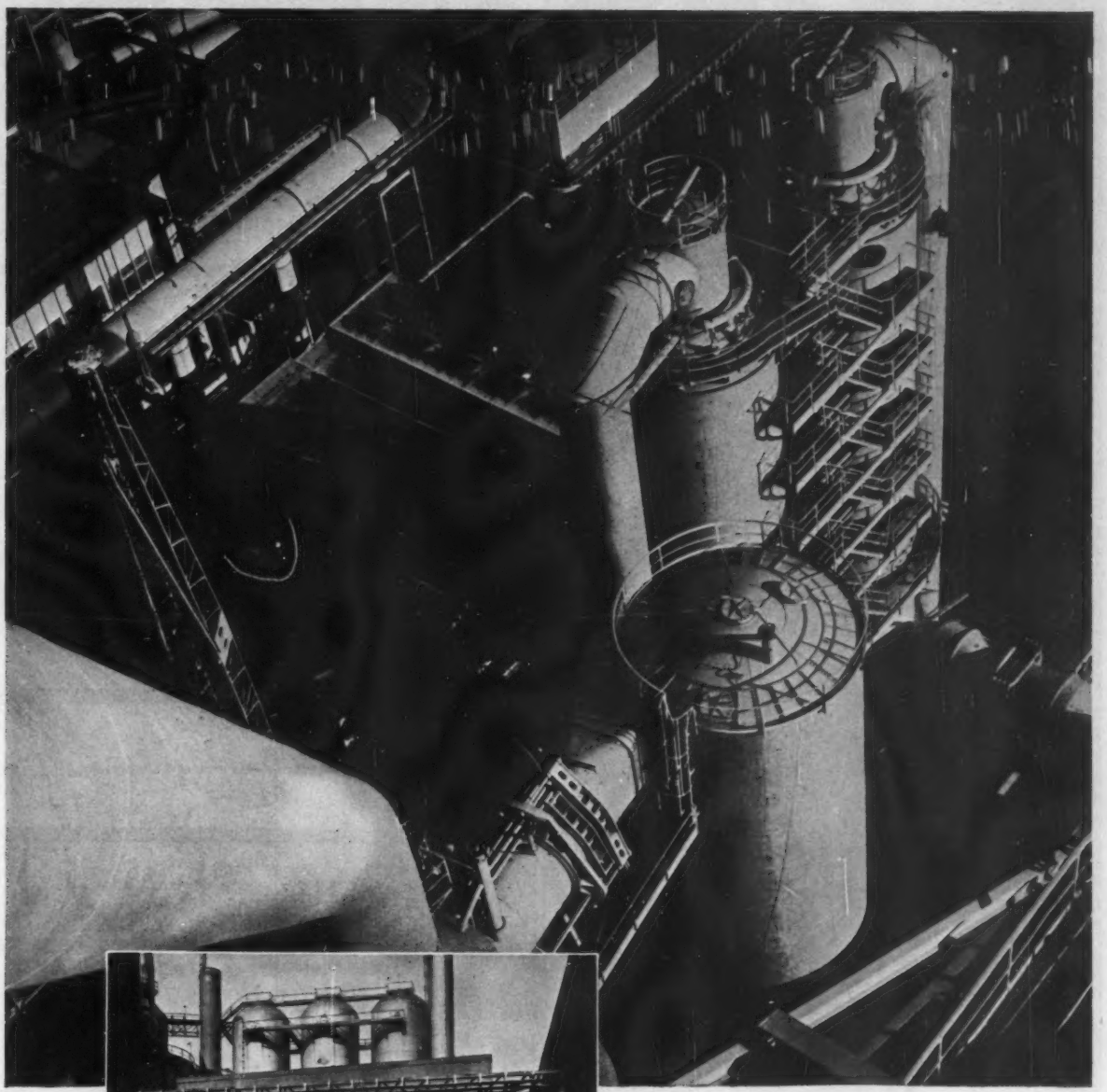

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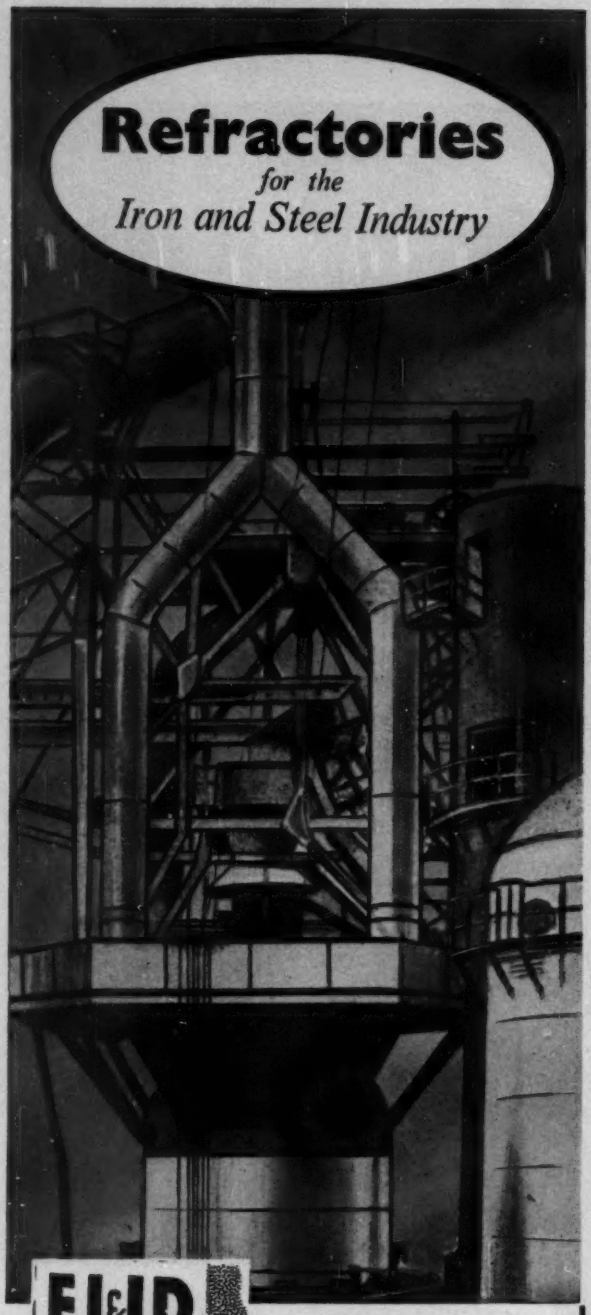

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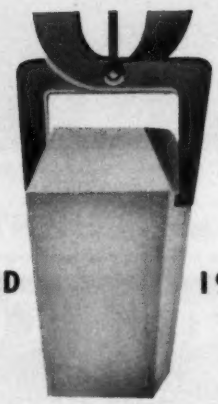

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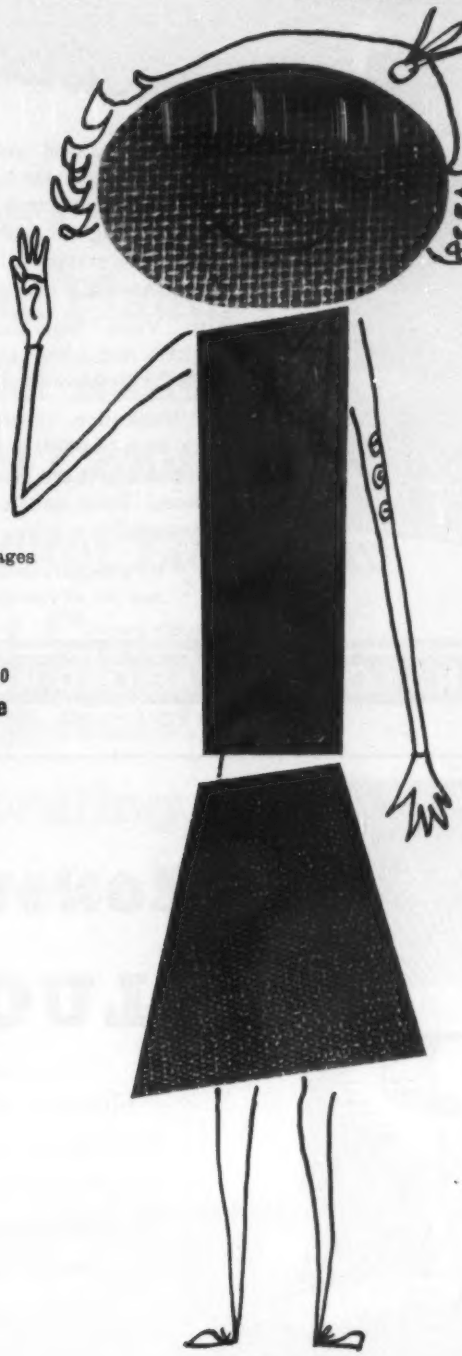

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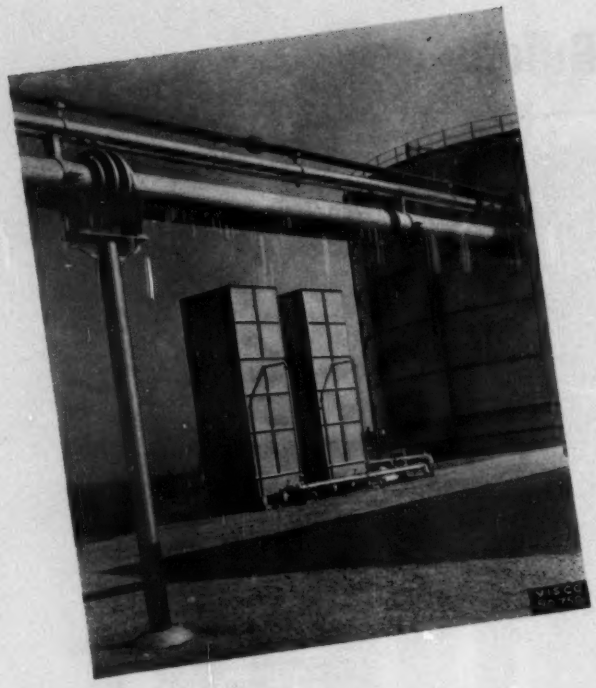

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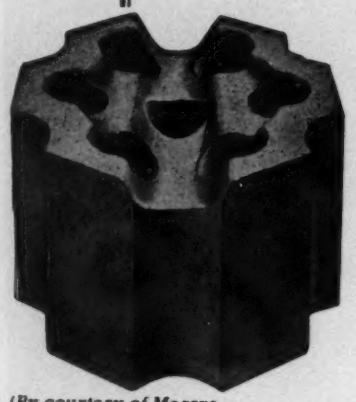

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THE IRON AND STEEL INSTITUTE

Special Report No. 60

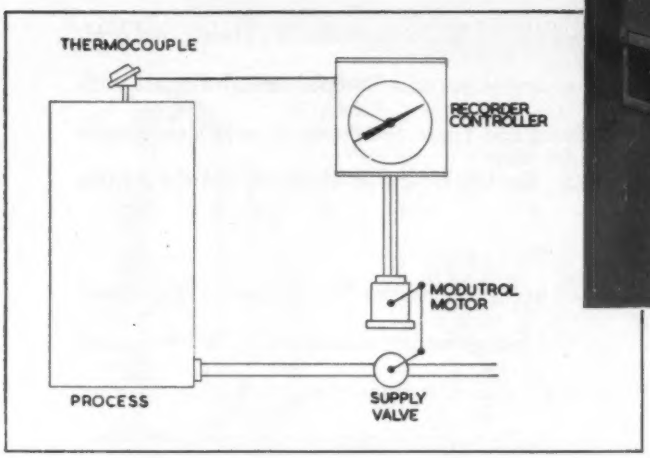
Proceedings of the

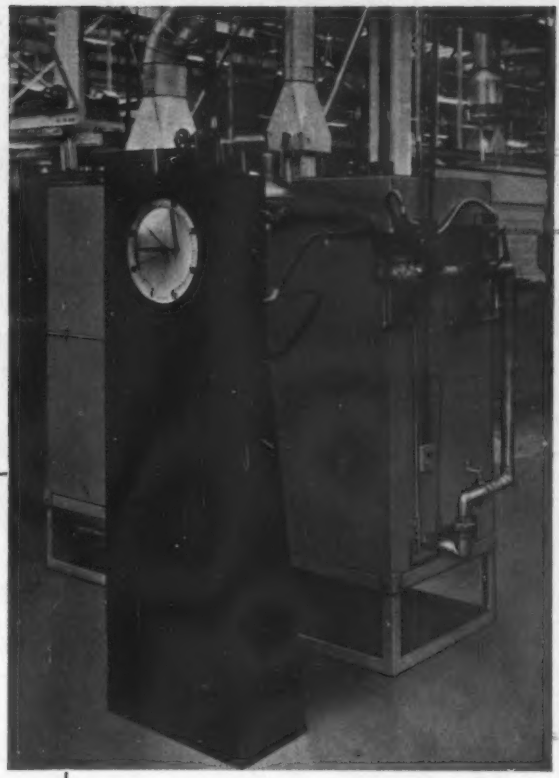
FORGEMASTERS' MEETING, 1954

In October, 1954, The Iron and Steel Institute held a Joint Meeting with the National Forgemasters' Association and La Chambre Syndicale de la Grosse Forge Française. Technical Sessions were held in London at which nine papers on forging theory and practice and forging plant were presented and discussed. The remainder of the Meeting comprised visits to the English Steel Corporation Ltd., Thos. Firth and John Brown Ltd., and Hadfield Ltd., in Sheffield.

The Proceedings at the Technical Sessions have been published as a Special Report. The five French papers and discussion are presented in English, but summaries of all the papers are also given in French. The Report, No. 60 in the Special Report series of the Institute, can be obtained from the Offices of the Institute, 4 Grosvenor Gardens, London, S.W.1, price 37*s.* 6*d.* (Members 25*s.*).

Vauxhall Motors choose Honeywell's Non-cycling Temperature Control System





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THE IRON AND STEEL INSTITUTE

Special Report No. 59

COMBUSTION AND HEAT TRANSFER IN AN OPEN-HEARTH FURNACE

A RECENT addition to the Special Report series of the Institute is entitled "Combustion and Heat Transfer in an Open-Hearth Furnace," by S. W. Pearson, M. W. Thring, and J. H. Chesters. This Report describes a large-scale trial carried out on a Templeborough furnace with a team consisting of groups from Messrs. Steel, Peech and Tozer, the Research and Development Department of the United Steel Companies Ltd., the University of Sheffield, and the British Iron and Steel Research Association.

The previous Templeborough trials, reported in Special Report No. 37 (1946), emphasized the necessity for aerodynamic studies, continuous roof-temperature measurement, and the maintenance of the highest practicable furnace pressure. The shift from producer gas to oil firing in the last ten years has added new objectives, e.g. the determination of optimum nozzle size, steam/oil ratio, and burner direction. The work of the International Flame Radiation Committee at Ijmuiden has given preliminary answers to some of the questions, whilst the theory of heat transfer initially expounded in Special Report No. 37 has been developed to a point where calculated melting times and fuel consumptions are of the same order as those found in practice. The main purpose of the present trial was to measure as many factors as possible, leaving only one free parameter in each verifiable part of the calculation; this could then be determined from the observed result and the value thus obtained used with some confidence in future calculations. A subsidiary objective was the comparison of flow pattern and mixing in model and furnace.

The Report, No. 59 in the series, is fully illustrated with diagrams and photographs and is bound in cloth with stiff board covers. Copies may be obtained on application to the Secretary, The Iron and Steel Institute, 4 Grosvenor Gardens, London, S.W.1, price 37*s.* 6*d.* (Members 25*s.*).

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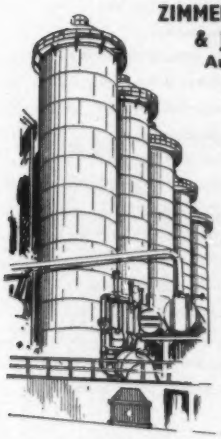

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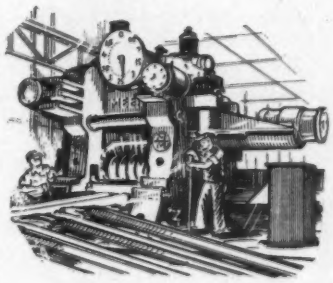

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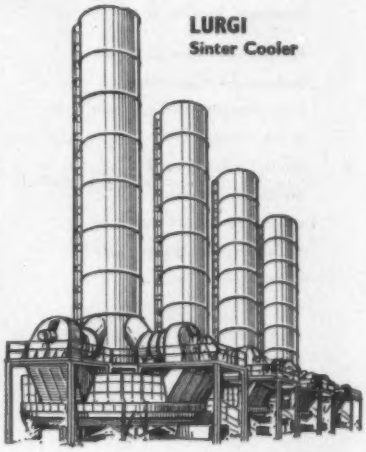

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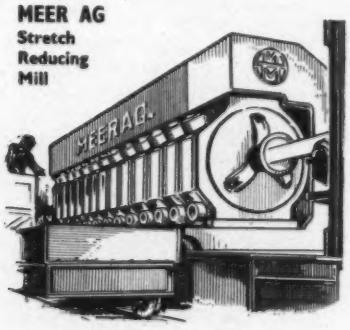

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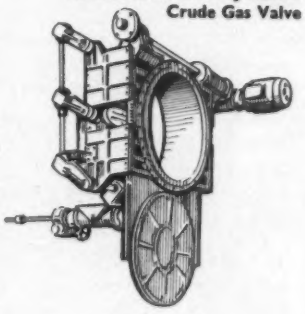

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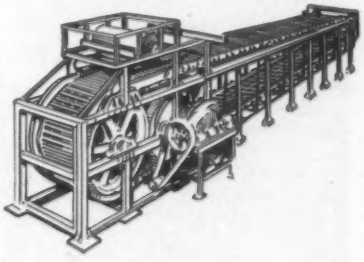

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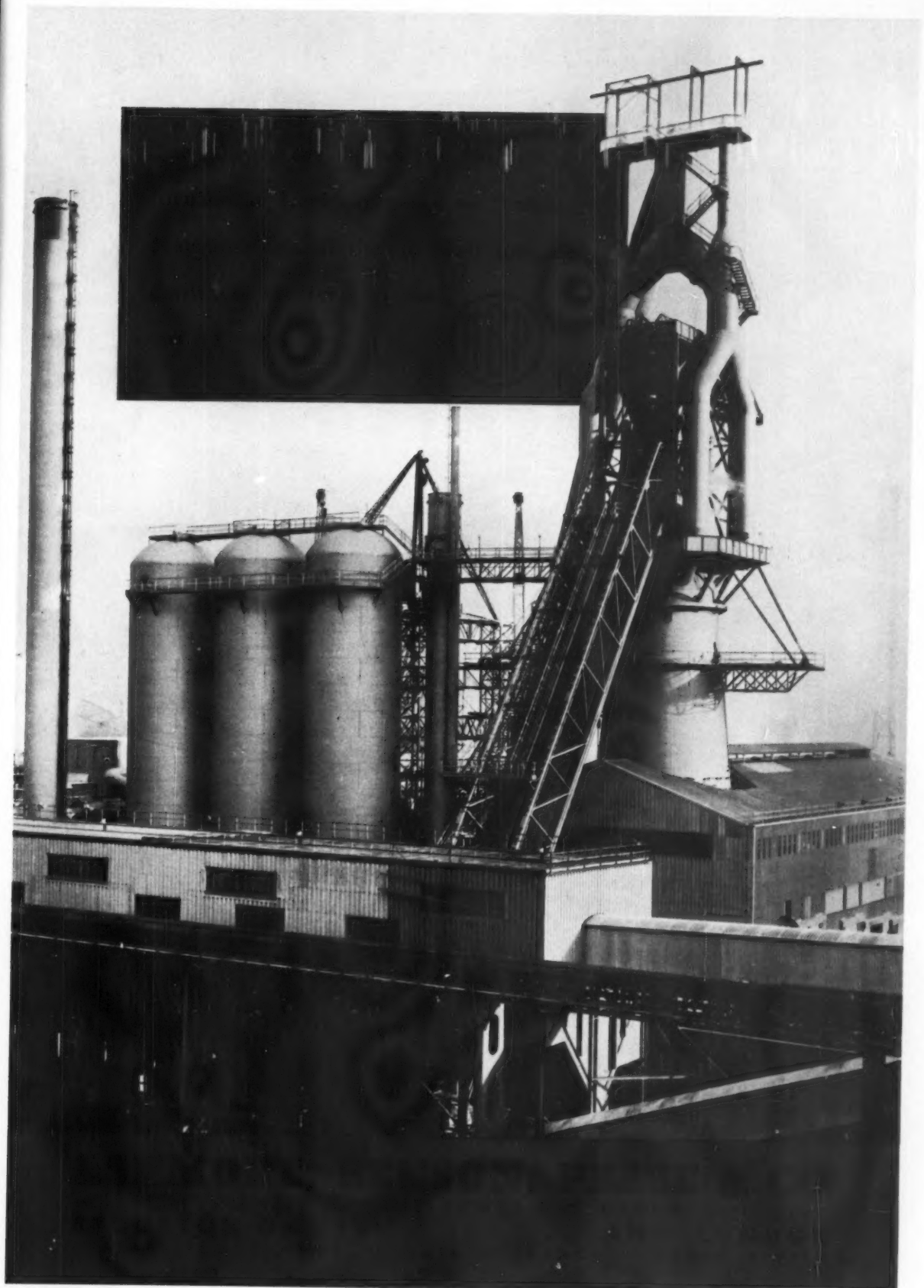

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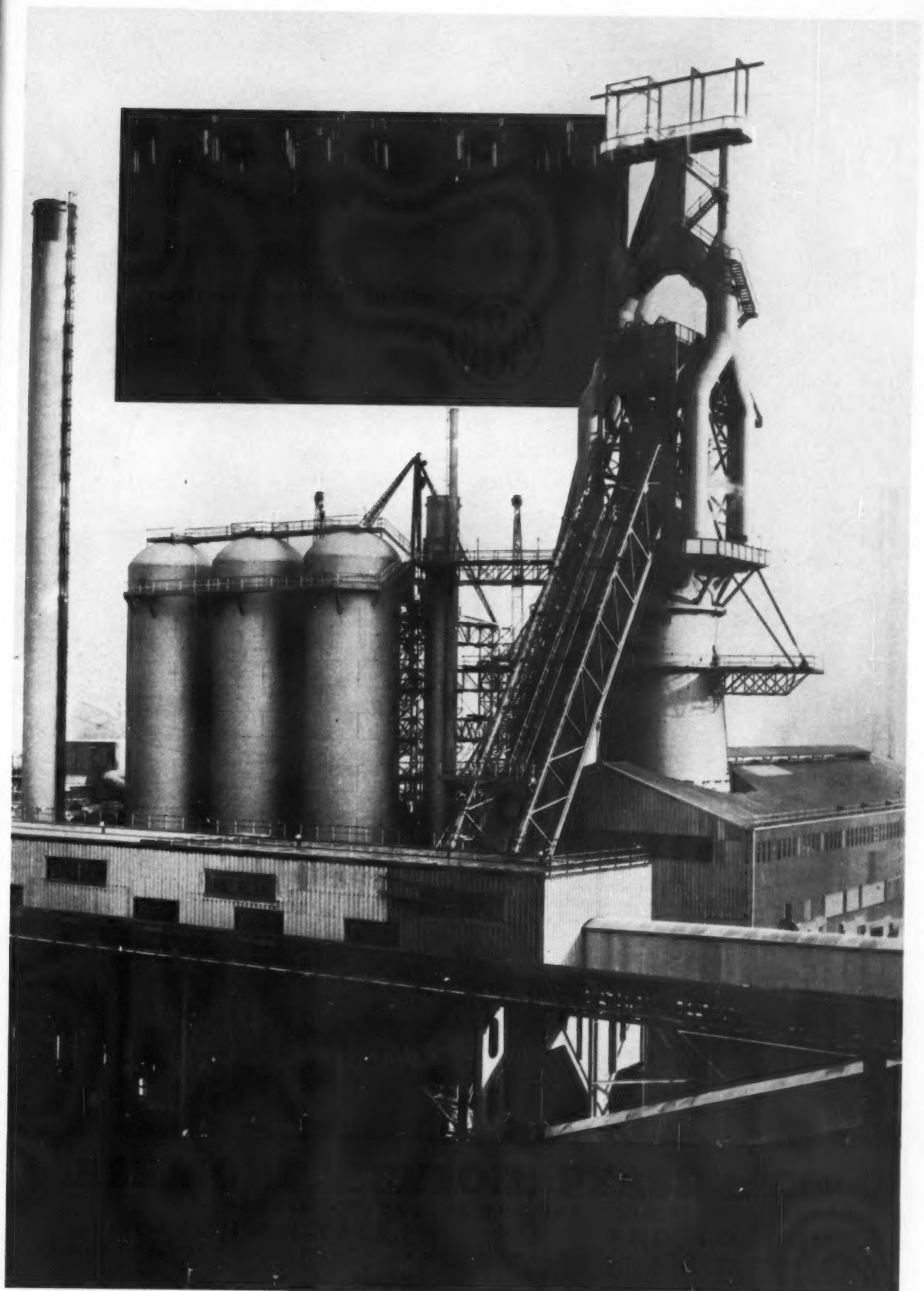

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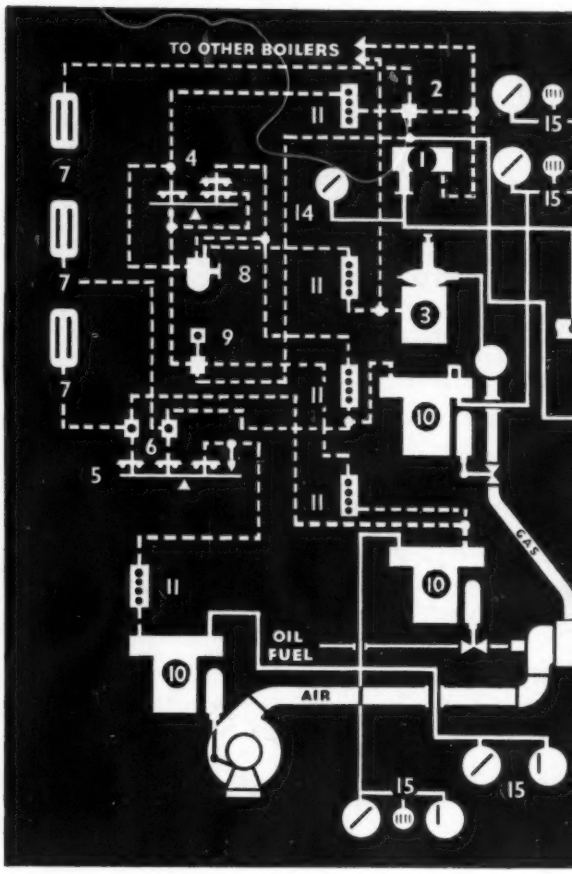

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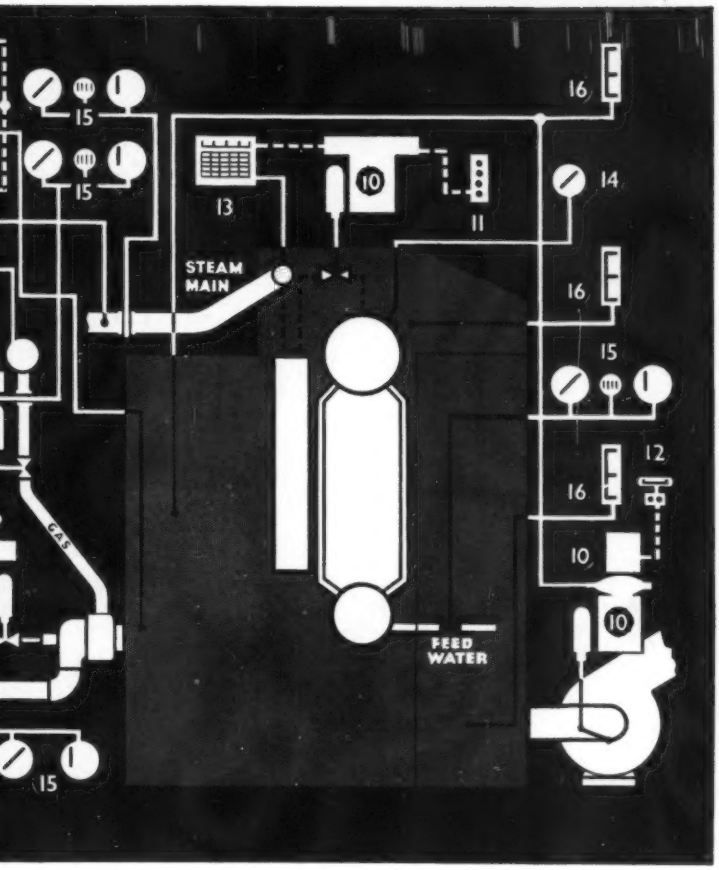

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